Transactions

AMERICAN FOUNDRYMEN'S ASSOCIATION

JAN 13 1941

Contents

Inverse Chill in Malleable Iron, *by Enrique Touceda*..... 449

Recent Experiments with Gray Iron Synthetic Molding Sands, *by Fulton Holtby and H. F. Scobie*..... 465

Chaplets and the Steel Casting, *by H. F. Taylor and E. A. Rominski* 481

Non-Ferrous Applications of Top-Pouring Methods, *by A. K. Higgins* 518

The Pearlite Interval in Gray Cast Iron, *by Alfred Boyles*.... 531

Effect of Manganese on Second-Stage Graphitization, *by D. P. Forbes, P. A. Paulson and G. K. Minert*..... 574

Influence of the Mold on Shrinkage in Ferrous Castings, *by H. L. Womochel and C. C. Sigerfoos*..... 591

Desulphurizing Cupola Cast Iron from the Practical Operating Angle, *by W. Levi*..... 623

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Inverse Chill in Malleable Iron

By Enrique Touceda*, Albany, N. Y.

Abstract

Investigations have been made on the subject of "inverse chill fracture" in malleable iron; but theories advanced seem weak, as methods of manufacture and structural and chemical composition are different in most experiments. Nine malleable iron test bar specimens were used in this investigation to see if causes of "inverse chill fracture," could be substantiated with theories outlined by other investigators. Metal samples were taken from the core, shell and intermediate areas of the test bars, if there was enough to conduct experiments with, tested and studied. These samples were subjected to chemical analysis, studied for defects, such as, gas holes, surface hard spots, entrapped gases and shrinks. Annealing and graphitization were investigated to study their effect on inverse chill fracture. These experiments did tend to show what the dominant facts in the inverse chill fracture of the various samples were. The investigation revealed that whether or not the fracture of a malleable iron casting will be normal, is wholly dependent upon the ductility of the ferritic grains in the casting. If ductility of the ferritic grains is lacking, it may be caused by a faulty anneal or by lack of metal density. The outside appearance of the test bars was thought to have been caused by the bleeding of the mold.

1. From time to time, inquiries have been made as to why, in certain instances in the testing of malleable iron tensile test bars, or in the breaking up of annealed castings, a fracture will be seen that is extremely abnormal. This particular fracture will have a white metal central core that may occupy a large or small part of the cross-sectional area. The core is surrounded by a shell, the fracture of which is not white. The appearance of the core corresponds to that of normally annealed malleable iron that can be seen in Fig. 1; specimens 1; 2; 3; 4; 6; 8 and 9.

* Malleable Founders' Society.

Note: This paper was presented at the 44th Annual A.F.A. Convention, Chicago, Ill. during the Malleable Iron Session, May 7, 1940.

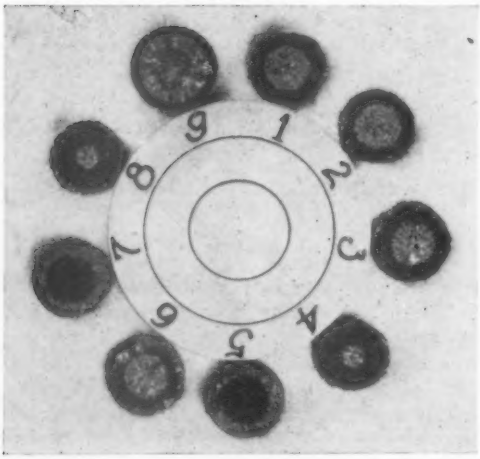

FIG. 1—NINE MALLEABLE IRON TEST BAR SPECIMENS SHOWING CROSS-SECTIONAL VIEW OF THEIR CORES AND SHELLS.

2. While the fracture of the shell and core of specimen 9 is, to the eye, similar in appearance to that of the other specimens, its structural and chemical make-up was found to be very different.

3. The fractures shown in specimens 5 and 7, Fig. 1, do not represent what are known as "inverse chill" fractures, for the metal of the shell is white and that of the core, black. They are fractures of deeply framed test bars which, under certain conditions of annealing, are at times produced, and are shown simply in order that the difference between these two types of fracture may be compared.

4. As far as the author was able to ascertain, the fractures of all of the castings, that were annealed at the same time as were the test bars that had an inverse chill fracture, were of normal appearance. Because of this fact, coupled with other reasons that will be set forth, it appears certain that the heat treatment to which the test bars were subjected was not accountable, except indirectly, for this character of defect. The author believes that this defect is due primarily to the particular conditions that prevailed during the freezing of the metal in the mold. Obviously, if this were not the case, the same type of fracture would be present in many, if not all, of the castings from the same air-furnace heat, annealed at the same time as were the defective test bars.

5. The author is of the opinion that this type of fracture has

been designated by the gray iron founders, who have written on this subject, as an "inverse chill," presumably because, when molten gray iron is cast against a chill, the fracture of the chilled part, to a certain depth, is white and the fracture of the rest of the cross-section is of normal appearance. Therefore, in a fracture where the reverse condition exists, they have simply designated it as an "inverse" chill, when obviously there could have been no chilling action on the metal of the core of the casting.

6. The author believes that in the case of either gray or malleable iron, the terms "inverse chill" is misleading and a misnomer. Various articles have been published, and explanations have been advanced, as to the reasons for the formation of inverse chill fractures in gray irons. The author is not aware that any investigations have been made as to the cause for this character of defect in malleable iron castings.

7. In experiments with gray iron some investigators have attributed this type of defect to (1) the presence of dissolved iron oxide; (2) to the fact that on cooling, the central part of the casting expanded against the solid shell and consequently was subjected to an abnormal amount of pressure during solidification; (3) to the fact that the pressure of the solidified shell on the central core was sufficient to prevent the formation of primary graphite; (4) to surface oxidation during pouring; and (5) to iron poured from a green ladle. Others have attributed the defect to segregation.

8. These facts are mentioned only as a matter of interest. Due to the difference in process of manufacture and the structural and chemical composition of these two products, what might prove to have been the cause in one case would not necessarily hold true in the case of the other, assuming that any one of the theories that have been advanced are tenable.

9. In studying the tensile bars showing an inverse chill fracture, the chemical and structural compositions were investigated. This was done to discover, if possible, whether these fractures were due to segregation, or to any of the other causes mentioned. In the analytical work on the test bar specimens 1; 2; 3; 4; 6; 8 and 9, Fig. 1, the results of which are recorded in Table 1, every precaution was taken to secure samples that truly represented the metal in the shell and core. Unfortunately, most of the specimens were of very limited size, being short pieces of the test bars only. In some cases they were not of sufficient bulk to secure a sample of

Table 1

CHEMICAL ANALYSES OF SAMPLES TAKEN FROM VARIOUS SECTIONS OF TEST BARS

Bar No.	*Sample From*	*Chemical Composition, Per Cent*				
		Si.	*P.*	*S.*	*Mg.*	*T.C.*
1	Shell	0.79	0.194	0.062	0.23	0.28
	Intermediate	8.84	0.225			0.78
	Core		0.211	0.059	0.26	0.82
2	Shell	0.95	0.169	0.071	0.27	0.49
	Core	0.95	0.167	0.068	0.28	1.03
3	Shell	0.98	0.169	0.074	0.27	0.79
	Intermediate		0.115			
	Core	0.98	0.275	0.057	0.29	0.85
4	Shell	0.88	0.198	0.068	0.26	0.76
	Intermediate		0.189	0.066		
	Core	0.92	0.209	0.066	0.24	0.79
6	Shell	0.95	0.160	0.076	0.28	0.45
	Intermediate	0.97	0.123	0.069	0.30	0.90
	Core	0.93	0.388	0.073	0.30	0.90
8	Shell	0.93	0.175	0.071	0.29	0.49
	Core	0.92	0.186	0.068	0.29	1.06

sufficient weight for some of the determinations. While great care was exercised in carrying through the analytical work, in only a few instances was there enough of a sample to do the work in duplicate.

10. In securing the drillings from the shell, care was taken not to encroach on the metal of the core. To be on the safe side, metal from the shell was removed only to a depth that closely approached the core. As a matter of interest only, where there was a sufficient amount of sample, an intermediate cut was taken that contained a small part of both shell and core, and finally metal was secured wholly from the core.

11. In Table 1, these different samples are marked "shell," "intermediate," and "core." In all of the samples, it was noted that, the metal in the core was found to be very unsound, which could easily be discerned by eye.

12. Specimens 1, 2, 3, 4, 6 and 8, in Fig. 2, are micrographs taken at the center of the specimen's cross-sectional area where unsoundness was most pronounced. These micrographs are num-

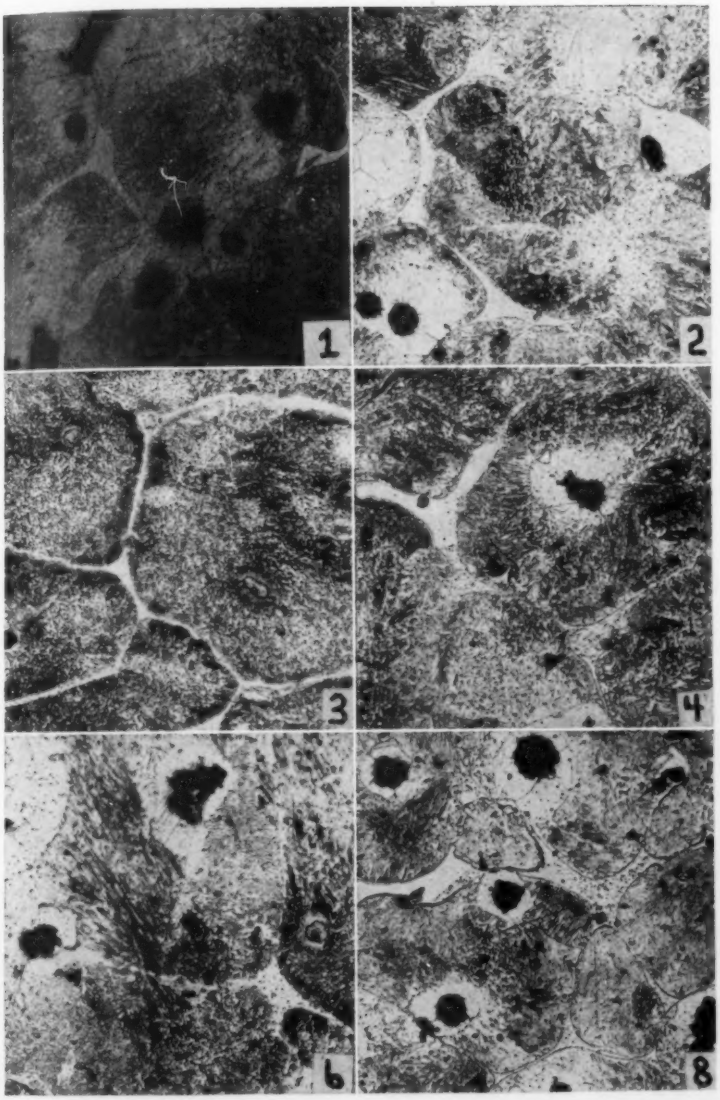

FIG. 2.—MICROGRAPHS TAKEN NEAR SPECIMENS CROSS-SECTIONAL AREA WHERE UNSOUNDNESS REGIONS WERE FOUND. SPECIMEN NUMBERS CORRESPOND TO THOSE IN FIG. 1. X100.

bered the same as are the specimens from which they were secured. (See Fig. 1.)

13. Dealing with the chemical composition of these various samples, it will be noted that there is a considerable amount of

phosphorus segregation in specimens 3 and 6, but not in specimens 1, 2, 4 and 8, and still the same type of fracture exists. This would indicate that while a serious amount of phosphorus segregation may have been a contributing factor in the case of samples 3 and 6, this is not essential to the formation of an inverse chill fracture.

14. It will be noted, also, that in specimen 1, the phosphorus in the intermediate area is higher than that in the core. The contrary is true in specimens 3, 4 and 6, where it is lower than in either the shell or the core. The author can offer no explanation, although in connection with segregation such vagaries are not uncommon.

15. A difference in carbon content between shell and core should be expected in an annealed casting. There is but a scant difference in manganese and silicon content in shell and core in any of the six samples, but there is a considerable difference in the sulphur content of specimen 3.

16. As previously indicated photomicrographs of specimens 1; 2; 3; 4; 6 and 8, Fig. 2, were taken at points towards the center of the cross-sectional area where the metal was most unsound. Fig. 3, 1-A, was taken in the shell of specimen 1, near the surface, and Fig. 3, 1-B, at a region just outside of the unsound part of

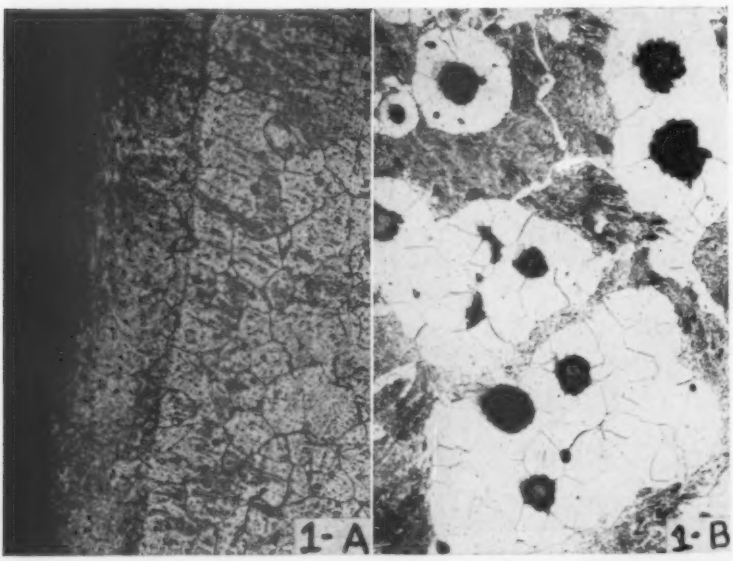

FIG. 3—MICROGRAPH 1-A (LEFT), REVEALS THE SHELL STRUCTURE OF SPECIMEN 1. 1-B SHOWS THE REGION OUTSIDE OF THE UNSOUND PART. x100.

specimen 1. These show the difference in structure between the respective areas. Not having the time to make similar micrographs of other samples, the following remarks will furnish a rough idea of the conditions that exist at these locations in other samples.

17. In some of these six specimens, there was a distinct parting line that separated a narrow surface layer from the metal directly beneath it. The exact nature of this parting line, due to lack of time, was not investigated. It probably would have been very difficult to have determined its exact character had time been available.

18. Such a parting line existed in specimens 1 and 2 about 1/64-in. below the surface. Between surface and parting line, the structure was very fine and consisted of pearlite.

19. Specimen 3 had no such parting line, but about 1/64-in. below the surface, the structure consisted of large, closely packed grains of pearlite. Specimen 4 had a distinct parting line about 1/64-in. below the surface, the structure between surface and this line being very mixed.

20. Specimen 6 had a distinct parting line, the structure between the surface and this line consisting of more or less fine crystalline grains of ferrite.

21. Specimen 8 had no parting line, but 1/64-in. below the surface, the structure consisted of very fine pearlitic particles that gradually melted into a fairly normal structure beneath it.

22. From the fact that in all of these six specimens there is a bull's eye structure, surrounding the unsound area, it can be concluded that the annealing cycle to which these samples had been subjected was not entirely correct. While the castings annealed at the same time did not have an inverse fracture, it is believed that, had they been examined, a bull's eye structure would have been found at some parts of the cross-sectional area. However, the author believes that the anneal only played an indirect part in the formation of the inverse fracture in these samples. Had the anneal been as perfect as possible, there still would have been an inverse fracture in the bars, due to unsoundness of the metal in the core, although an imperfect anneal obviously would result in a worse condition than if the anneal had been perfect.

23. Inasmuch as the metal in the core of these test bars was

unsound, it follows that during the freezing of the metal in the mold, the shell evidently did not exert a pressure on the metal of the core. Consequently, pressure was not a factor in the formation of an inverse chill fracture in the case of these specimens.

24. In view of the fact that the metal in the core is unsound, the presence of particles of free cementite should be expected. Many years ago, in an investigation by the author as to the cause for hard spots in a casting, he ascertained that where hard spots existed there was a shrink. If there was not a shrink, there was at least some incipient unsoundness in the metal. Where either of these conditions existed, there invariably were particles of free cementite, a considerable amount in the case of the former and a lesser amount in the latter, that is, the more serious the unsoundness, the greater the number of particles of free cementite.

25. Further study on the subject investigated showed that, irrespective of the character of an annealing schedule, graphitization seriously was interfered with at locations where the metal lacked normal density.

26. That a dense metal can be graphitized in a shorter time than one less dense was demonstrated by Forbes. He found on rolling down some small hard iron slabs into sheets at a rolling temperature, that when the sheets subsequently were annealed, graphitization took place in a fraction of the time required to anneal castings from the same heat. This confirms the fact that, as far as graphitization is concerned, the density of the metal is a factor.

27. Surface hard spots are caused in the same manner. That is they can be caused by pellets of fines present in the heap sand. The pellets contain an abnormal amount of water, and if they, or a piece of core butt that contains some unburned core-compound, happen to be located at the face of a mold, the steam from the former, or the gases from the latter, when suddenly released, will disturb the solidifying crust of the metal where they are located. Thus, small areas of unsound metal are created, and complete graphitization will not take place at these areas.

28. Unsound metal also can be caused by entrapped gases in the metal, particularly if it has been poured at too low a temperature. When the molten metal is dull, the gases cannot escape. This may account for the unsoundness in test bars 1; 2; 3; 4; 6

and 8, and for the formation of the inverse chill fractures in the case of these particular specimens. An important factor against this conclusion is the fact that the minute cavities do not look like gas holes. On the other hand the surface of these bars showed no wavy appearance or any evidence that the pouring temperature had been abnormally low. Inasmuch as the cavities do not look like gas holes, it might easily be that the unsoundness resulted from the fact that the molds may have bled, to some extent, and the minute cavities have been formed because of this fact.

29. In connection with specimen 9, the author should explain that a very short piece of this test bar was found among a lot of freak-fractured specimens that the author had put aside in the past, awaiting an opportunity to examine them. This test bar had the same character of fracture as the specimens thus far considered, and it was examined as part of a malleable iron test bar. Unfortunately no record was attached to the specimen regarding its character or history. The piece was so short that there was but a scant amount of metal available for chemical analysis, and at the start of the work, and in ignorance that the sample contained 14.60 per cent silicon, great difficulty was experienced in getting the metal into solution with the usual solvents.

30. The material in the shell was so soft that it easily was removed, in the form of very fine powder, by a file. The steel was so soft that it was found impractical to make a microscopical examination as the material would not take a polish. Only a sufficient amount of sample was available for chemical analysis of the elements shown in Table 2. When a portion of the powder was heated in an alkaline solution, an odor of ammonia, or of some organic base, appeared to be noticeable. For that reason, a nitrogen determination was made. Only enough of the material was available for metallic iron; silicon; copper and nitrogen determinations. The

Table 2

PARTIAL ANALYSIS OF SHELL AND CORE FOR SPECIMEN 9

Shell		*Core*	
Element	*Per Cent*	*Element*	*Per Cent*
Metallic Iron	6.60	Silicon	14.60
Silicon	17.00	Carbon	0.88
Copper	1.15	Manganese	0.62
Nitrogen	0.10	Copper	2.08
		Chromium	0.098

metal of the core was very hard (C-48 Rockwell) and very brittle, and there was just enough of the sample to make the chemical analyses shown in Table 2.

31. A microscopical examination of the core, which was more or less unsound, showed the barley-like figures that are seen in an etched sample of a high silicon (14-15 per cent) iron, and that this test bar was cast from a heat of such an iron, seems to be obvious.

32. The final specimen to be considered, is a heavy chain link containing an inverse chill fracture in the side-member. The chill was somewhat less than 1 sq. in. in cross section. A polished and etched cross-section, at a magnification of about 3 diameters, is shown in Fig. 4. This shows that the fracture of the metal of the core, while of a wholly different appearance than that of the normal metal surrounding it, is steely and not white, as was the case in the other samples.

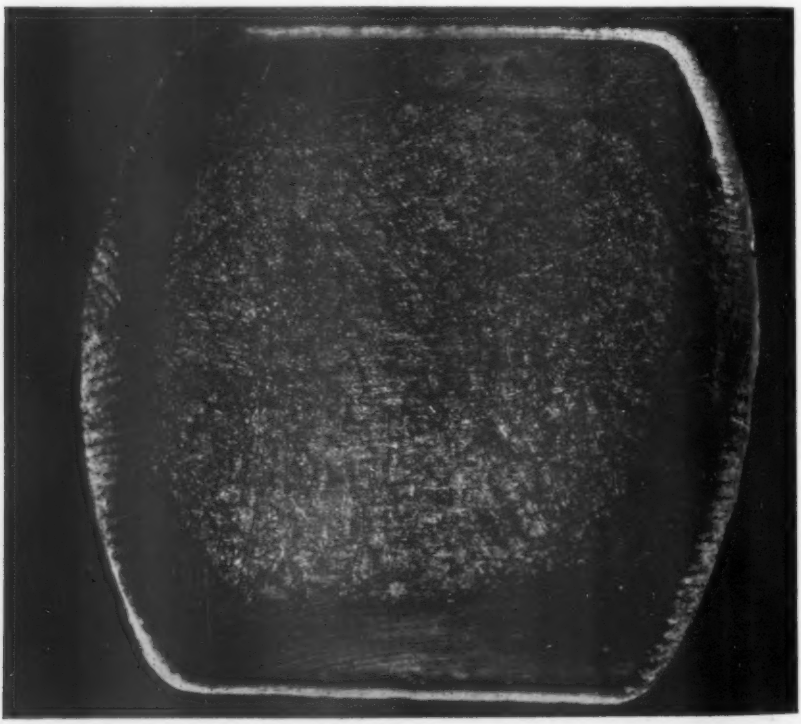

FIG. 4—CROSS-SECTIONAL VIEW OF CHILL AREA FOUND IN A CHAIN LINK. x3.

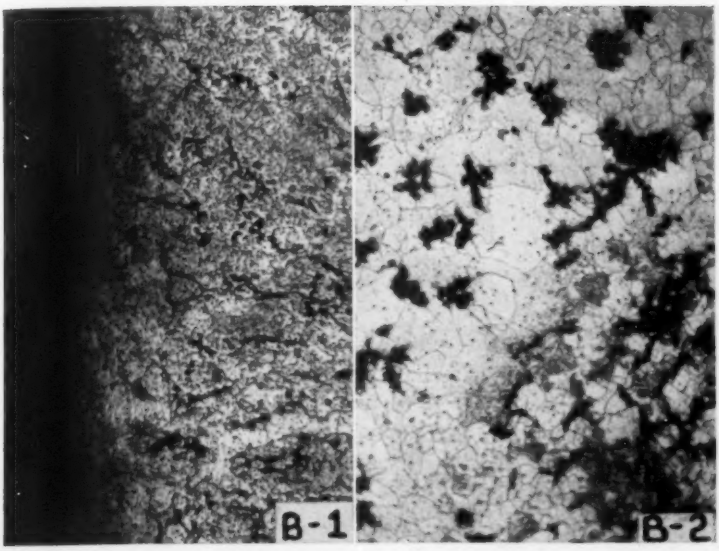

FIG. 5—MICROGRAPHS OF CHILL SHOWN IN FIG. 4. B-1 (LEFT) REVEALS STRUCTURE OF THE SHELL AND B-2 THE STRUCTURE OF THE CORE.

33. When the sample was fractured at different places, within a distance of several inches to each side of the inverse chill, this type of fracture gradually disappeared. The shell became progressively thicker and the core smaller until the fracture became normal throughout the entire cross-section, that is, the inverse chill was confined to only a certain length of the side member.

34. Fig. 5, B-1 and B-2, show respectively, the structure of the shell and core. A chemical analysis of each of these parts was made and is shown in Table 3.

Table 3

CHEMICAL ANALYSIS OF SAMPLE SHOWING INVERSE CHILL FRACTURE

	Si.	*P.*	*S.*	*Mg.*	*T.C.*
	%	%	%	%	%
Shell	1.17	0.130	0.035	0.280	1.62
Core	1.13	0.130	0.131	0.280	2.27

35. It will be noted that the sulphur content of the core is almost four times that in the shell. A difference in the carbon content would be anticipated, due to this condition.

36. Before attempting to draw conclusions as seem pertinent in connection with this investigation, it would be well to briefly

enter into a discussion of the various causes that operate to produce a sparkling, a steely, or a white fracture in a test bar or other malleable casting.

37. The fracture of any casting of well annealed malleable iron invariably is black or grayish in color, depending upon the angle in which the light strikes the plane of the fracture. The dark color results from the fact that the crystalline grains of ferrite are so ductile that, when a casting is broken in tension, they elongate into spines and it is the shadowy region between them that is accountable for the dark color of the fracture. Therefore, anything that operates to lessen the ductility of the crystalline grains in the matrix of a malleable iron casting, will lessen the length of the spines. Consequently, the shadows between them will be less concentrated, with the result that the fracture will take on a whitish color. The shorter the spines, the whiter will be the fracture. If ductility is wholly lacking in the entire cross-sectional area, or at parts only, the section will be wholly white in the former case and only steely or white at certain spots in the latter case.

38. This becomes apparent when a piece of malleable iron is fractured by bending. The extreme fibres on one side will be in tension while fibres on the other side will be in compression. Because of this fact the fracture on the compression side will be steely in color, as the grains on that side have been flattened, while those on the other side will have been elongated and the fracture will be normal.

39. A sparkling fracture can be caused by the presence of isolated patches of pearlite, or by particles of free cementite irregularly or more or less uniformly distributed throughout the matrix. These break short and reflect light, while if these particles are more numerous, the fracture may be steely throughout.

40. Obviously, ductility can be affected adversely in a number of other ways. For example, a piece of well annealed iron can be broken so suddenly that the elongation of the ferritic grains is denied and, consequently, the fracture will be whitish or steely.

41. As has been explained, where any unsoundness exists in a casting, there will be a more or less amount of both pearlite and free cementite in that particular area, depending upon the degree of unsoundness and its extent.

42. In photomicrographs of specimens 1, 2, 3, 4, 6 and 8, Fig. 2, numerous particles of massive free cementite can be seen at the boundaries of large pearlitic grains. Because of the presence of these coarse grains of pearlite and the large free cementite particles, it could have been predicted that the color of the fracture, where these areas occur, would be white.

43. In conclusion, the experiments have shown what are to be considered as the dominant facts in the inverse chill fractures of the various samples. Whether the fracture of a malleable iron casting will be normal, steely at certain locations, steely throughout, white at certain locations, or white throughout, is, in the author's opinion, dependent wholly upon the ductility of the ferritic grains throughout the entire section, or at certain parts only, unless, as has been pointed out, the casting is fractured with great suddenness.

44. A lack of ductility of the ferritic grains may be affected through a faulty anneal as well as by a lack of metal density.

45. Because of the outside appearance of the test bars 1, 2, 3, 4, 6 and 8, the author is inclined to believe that a bleeding of the mold was the most probable cause for the unsoundness in these test bars.

DISCUSSION

Presiding: C. C. LAWSON, Wagner Malleable Iron Co., Decatur, Ill.

E. K. SMITH:[1] I recently read an article by a British metallurgist concerning this matter of inverse chill in gray iron. He had studied many samples of inverse chill and had never found one that was not excessively high in sulphur, particularly in the chill. That checks in with Dr. Touceda's findings on the malleable.

PROF. TOUCEDA: You are referring to gray iron, and what would hold true in the case of gray iron would have no bearing in the case of malleable.

MR. SMITH: You find a higher sulphur there and they found the same thing in the gray iron. It might not have any bearing but it would seem to be comparable.

PROF. TOUCEDA: You are referring to but one of the cases cited which was not the situation in the rest of the samples, in which cases the sulphur was normal throughout. It was only in the case of the link where sulphur segregation had taken place.

MR. SMITH: In the gray iron, this British author said that there were many different conditions. He mentioned high pressure, but he did

[1] Metallurgist, Electro Metallurgical Corp., Detroit, Mich.

remark on high sulphur. He had no particular theory but he said he had never found one with that inverse chill unless it was connected with a high sulphur.

R. J. COWAN:[2] What was the form of these test pieces?

PROF. TOUCEDA: They were the regular ⅝-in. standard test bars.

MR. COWAN: Have you ever found that condition in a low-silicon, low-carbon iron? I noticed the silicon percentage was 0.98. Suppose the percentage of silicon had been 0.85 and the carbon 2.40? Would you find an inverse chill in a low carbon, low silicon iron of that type?

PROF. TOUCEDA: I do not think so. In the early days, that is the type of material that was made by all of the plants and, during this interval, the speaker had never heard of an inverse chill, although there may have been an occasional one.

MR. COWAN: Do you find inverse chill in castings?

PROF. TOUCEDA: No, I have never found them in a casting other than in a test bar except in the case of the link to which reference has been made. We are continually testing bars and when we find one with this type of fracture it is set aside for examination.

MR. COWAN: I was very much interested in the fractures illustrated. Is it possible, by means of heat treatment, to reproduce any of those?

PROF. TOUCEDA: No, I do not think so. I know of no heat-treatment by which that could be accomplished. How could one heat-treat a material of this kind locally to give the outside one treatment to produce the structure in the shell and the inside another treatment to produce the structure in the core. I am quite certain that this could not be done. I do not think heat-treatment has anything to do with this case any more than it has to do with a shrink.

MR. COWAN: I have observed castings that showed this phenomenon of an inverse chill. I was never able to carry that throughout the entire casting. That is to say, an examination would show that it existed in heavy sections but not in light sections of the same casting, and always in heavy sections that had been subsequently re-heated as a result of their position in the mold. As a result of various heat treatments we have carried on in our laboratory, I believe I could reproduce any one of those fractures. I have seen those structures reproduced as a result of various heat treatments. I am just wondering if this may not be a result of some secondary re-heating as a result of the location in the mold, with a high-silicon, high-carbon iron.

PROF. TOUCEDA: I do not think that you can produce these structures by any type of heat-treatment. They are the result of what happens to the metal after it is in the mold. I do not see how you could have any selective treatment in a casting or in a bar to produce this type of fracture.

[2] Metallurgist, Surface Combustion Corp., Toledo, O.

C. F. JOSEPH:[3] I have never heard of any inverse chills in malleable iron before. We have made test bars for a period of years and I have never seen any of them that had this inverse chill. I have seen rimmed bars, framed bars, but that is not what you are talking about.

PROF. TOUCEDA: No, I just showed two fractures that illustrated framed bars for the sake of comparison with the other.

MR. JOSEPH: As far as malleable castings are concerned, I have never seen any which showed this white rim or this white inverse chill that you are mentioning, so I do not believe it is very common at all.

H. BORNSTEIN:[4] I am in a similar position to Mr. Joseph in that I think Prof. Touceda has the right answer. I think it is tied up with segregation, something like is found in the shrink area. The structures are somewhat similar to what is found in a shrink area where there is not complete graphitization, and I can conceive that this may be an accentuated shrink area.

PROF. TOUCEDA: I believe that the trouble results from a bled mold. I do not think it has anything to do with the use of a green ladle; nor with dissolved oxide; nor do I think it has anything whatever to do with the pressure of the metal shell on the core, because if that were true, the core would not be unsound, and in the cases cited all are unsound. Some gray iron men who have written on this subject suggest that instead of the shell metal pressing on the core, the metal in the core expands and presses against the metal in the shell, but this action would be the same as the other. I am not certain that I am absolutely correct in my opinion in regard to the cause being laid wholly to bled molds, but just prior to leaving for the Convention, a test bar had been received where an attempt had purposely been made to bleed the mold. In this case, there was not anywhere nearly as much free cementite in the porous regions, but it did have very large grains of pearlite.

CHAIRMAN LAWSON: Does the author believe that shrinkage conditions, even if the mold had not been bled, would cause this condition?

PROF. TOUCEDA: No, I do not. There is a typical structure in a shrinkage area and, if the shrink is very light, it will consist of fine grains of pearlite, but if the shrink is rather serious, there will be considerable cementite present. In the cases shown, however, the grains of free cementite are very massive and so are the grains of pearlite.

CHAIRMAN LAWSON: Would not a bled mold have to show some shrinkage or porosity?

PROF. TOUCEDA: Mr. Bornstein gave an illuminating answer to this question. He stated that the condition represented a very large shrink. Well, this is just what it is. It is a shrink covering a rather large volume of metal. In other words, if we could have taken these bars and deter-

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[3] Metallurgist, Saginaw Malleable Iron Div., Saginaw, Mich.
[4] Director of Research, Deere & Co., Moline, Ill.

mined the density of the shell metal and the density of the core, that would have been the answer. But we could not get samples to do that and it is very difficult to get the specific gravity of a sample that is unsound. The reason that the paper was prepared was, I thought it was more or less novel and new and it was something that probably might bring forth certain investigations and we might eventually learn the truth by someone else investigating them. I think that is interesting, too, that you can produce this result by heat treatment. Personally, I am almost certain that it is impossible, but I would like to have the effort made and I would like to get a sample if they do try it.

MR. COWAN: I had an experience some time ago with a certain casting that was being made and every once in a while we would have trouble, always on that one casting. We got this very phenomenon that Prof. Touceda is speaking about. It would always occur in the heavy sections of the casting, always when the metal was poured cold and when the silicon and carbon were higher than usual. We attributed it at the time to a secondary heating as a result of the mold staying in there and solidifying as it got a kind of partial anneal and, being high-carbon, high-silicon iron, it just quickly annealed at the edge. That all ties in with shrinkage too, because the same sort of thing happens in a shrink.

PROF. TOUCEDA: Anything that will interfere with the ductility of the constituents in the matrix will have a great influence on the character of fracture, and that is my whole story. In the case of gray iron, the situation is practically the same in that if, due to any cause, the combined carbon is greater in the core than in the shell, the ductility of the metal in the core is less.

MR. JOSEPH: I might add that we did add a carbide stabilizer in very small amounts and raised the silicon content from 1.25 to 1.50 and the gate end of the test bar did show a very small white area in the core, but it was not very large, due to the fact, possibly, that the carbide stabilizer was added in too small amounts. The heavy end of the test bar did show the characteristic gray around the outside before the bar was malleableized. After the bar was malleableized, it still showed the white core but it was a very, very small area of white. I do not know whether there would be anything in the fact that there might be small amounts of carbide stabilizer present but in this particular case we did add a very small amount, 0.003 or 0.004 per cent, of a certain carbide stabilizer to prevent the iron from mottling. Apparently we did not add enough of it to produce it all white.

Recent Experiments with Gray Iron Synthetic Molding Sands

By Fulton Holtby* and H. F. Scobie*, Minneapolis, Minn.

Abstract

This paper is a progress report of some work on synthetic sand mixtures being conducted by the authors at the University of Minnesota, Minneapolis, Minn. The fore part of the paper is devoted to a discussion of the occurrence and properties of the St. Peter's sandstone formation found throughout the mid-western states, with particular reference to the sand of that type which occurs in Minnesota. The second section of the paper deals with the utilization of steel sand fines in the molding of gray iron castings. These so-called steel sand fines include the sand and other materials removed from the sand used in a steel foundry to control the sand's properties. In this investigation, the authors found that steel sand fines can be used as molding sands of low permeability and as blending sands to control the properties of molding sands of similar composition. The authors also have studied the rapid drying of synthetic molding sands and have experimented with various substances as wetting agents as substitutes for water to avoid this difficulty. A small amount of information on this subject is given at the end of the paper, stating the results of preliminary tests.

1. Natural bonded iron molding sands commercially available in Minnesota are generally low in permeability, sintering point, and durability. They are usable, but cause trouble if constant sand control is not practiced. The Foundry Department, University of Minnesota, realizing the shortage of good local molding sand, is conducting a series of tests using synthetic molding sand for gray-iron castings. Only the preliminary experiments which have been completed will be discussed in this paper, which is a progress report.

* Instructor, Foundry Practice, University of Minnesota.

Note: This paper was presented at the Sand Research Session, 44th Annual A.F.A. Convention, May 9, 1940, Chicago, Ill.

SYNTHETIC GRAY IRON MOLDING SAND MADE FROM ST. PETER SANDSTONE

2. The St. Peter sandstone formation, consisting of nearly pure silica and nearly uniform composition, underlies a large area in Minnesota, Wisconsin, Illinois, Iowa, Missouri, and Arkansas, as shown in Fig. 1. The material is a friable sandstone consisting mainly of well-sorted quartz grains. The average thickness of the formation for the greater part of its distribution is about 100 ft.

3. In a recent study of this formation, Dr. George A. Thiel*[1] found that, on the average, the sandstone of the Illinois district is coarsest, the Arkansas district is somewhat finer, and the Minnesota district yields the finest grain. The slight variation in grain size occurs both horizontally and vertically, with the upper 10 to 20 ft. of the formation usually coarser than the central portion.

4. The grain shape and appearance of the St. Peter sand available in Minneapolis is shown in Fig. 2.

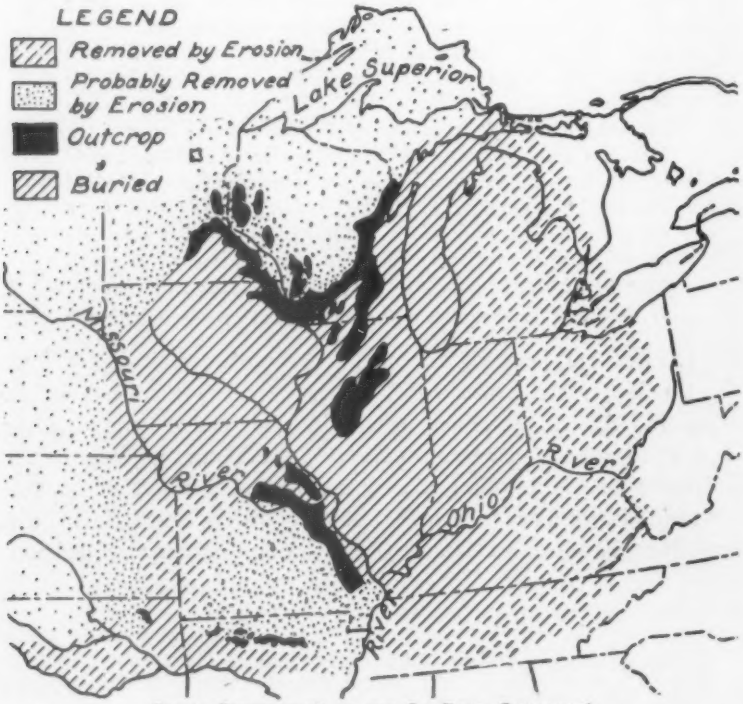

FIG. 1—DISTRIBUTION OF THE ST. PETER SANDSTONE[1].

* Superior Numbers refer to bibliography at end of this paper.

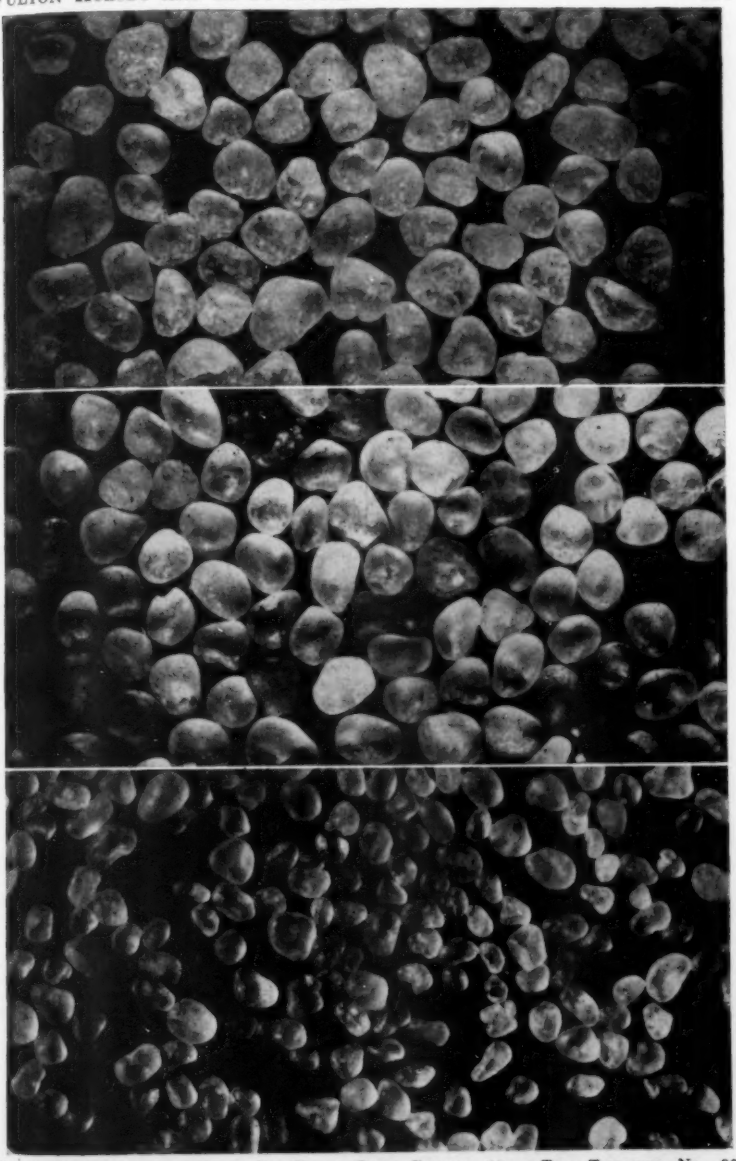

FIG. 2—PHOTOMICROGRAPHS OF ST. PETER SAND GRAINS, 12X. TOP—THROUGH NO. 20, RETAINED NO. 40 TYLER SCREEN. CENTER—THROUGH NO. 40, RETAINED NO. 70 TYLER SCREEN. BOTTOM—THROUGH NO. 70, RETAINED NO. 140 TYLER SCREEN.

5. For detailed information as to the texture of this sandstone for any particular location, reference should be made to Dr. Thiel's report.

6. This material is readily available in the Twin Cities area. It is used by the Ford Motor Company in the manufacture of glass and by local foundries for core sand. The construction of a series of tunnels for sewer pipes under Minneapolis and St. Paul resulted in approximately 400,000 cu. yd. of this sand being piled as waste material during the past 5 years. One pile, located within one-half mile of the University of Minnesota, offers a ready supply for research purposes. The analysis of this sand is shown in Table 1.

Table 1

SCREEN ANALYSIS OF ST. PETER SANDSTONE MINED HYDRAULICALLY AT MINNEAPOLIS

Screen Number	*Per Cent of Grains*	
	Sample 1	Sample 2
20	0.0	0.0
30	0.2	0.2
40	5.4	4.9
50	32.8	31.1
70	21.3	21.9
100	16.3	16.9
140	18.0	19.9
200	5.9	5.0
270	0.1	0.1
Pan	0.0	0.0
Bond	0.0	0.0

7. Various mixtures of this sand, bentonite, and cereal binders were made and tested. The composition and properties of what we call No. 10 mixture, is shown in Table 2.

8. This mixture molds well and has been used to make molds for castings weighing up to 150 lb. It is exceptionally good for machine-made test bar molds, such as flowability tests and step bar tests, as duplicate molds can be produced.

9. This synthetic molding sand, after shake-out, can be conditioned in a gyratory riddle. Our standard practice is to rebond the sand after every three heats to maintain a high compression strength. The proportions of materials used are shown in Table 3.

10. This synthetic sand had decreased in grain size, after about 75 heats, as shown in Table 4.

Table 2

COMPOSITION AND PROPERTIES NO. 10 MIXTURE
SYNTHETIC MOLDING SAND

Composition by Volume

24 parts new St. Peter Sand
¾ part Bentonite
¼ part Cereal Binder
1½ parts Water

Dry ingredients mulled 3-min., water added and wet mixture mulled 5-min.

Properties A.F.A. Standard Tests

Water, per cent	3.5
Green Permeability	80
Green Compression, lb. per sq. in.	5
Sintering Point, °F.	2507 ± 10
Fineness Number	63.8

Table 3

COMPOSITION AND PROPERTIES OF REBONDED NO. 10 SAND MIXTURE

Composition by Volume

24 parts used No. 10 Mixture
⅛ part Bentonite
⅛ part Cereal Binder
1½ parts Water

Dry ingredients mulled 3-min., water added and wet mixture mulled 5-min.

Properties A.F.A. Standard Tests

Water, per cent	3.3
Green Permeability	84
Green Compression, lb. per sq. in.	5
Sintering Point, °F.	2550 ± 10

11. Although heavy castings have not been poured in this mixture, the difference in the grain texture of the sand before and after 75 heats, as shown in Tables 1 and 4, indicates that it has good durability.

12. The permeability of the sand seems to improve with use. The increase in permeability can be accounted for by the large grains' decreasing to more nearly the size of the average grain. The increase in sintering point is probably due to an accumulation of cereal flour which would tend to burn to gas when the mold was poured. Sea coal facings have been used with other floors of this synthetic sand to raise the sintering point of the sand and to improve the surface of castings poured at 2800°F.

Table 4

SCREEN ANALYSIS AND PROPERTIES OF SYNTHETIC MOLDING SAND NO. 10 USED 75 HEATS, REBONDED 25 TIMES

Screen Number	*Per Cent of Grains*
20	0.2
30	0.4
40	8.2
50	16.8
70	31.8
100	18.6
140	15.4
200	4.2
270	0.4
Pan	0.6
Per Cent of Bond	3.3
Total, per cent	100.0
Water, per cent	2.5
Green Permeability	104.0
Green Compression, lb. per sq. in.	5.5
Sintering Point, °F.	2600 $\pm$ 10
Grain Fineness Number	64.2

13. The bad features of this synthetic sand mixture are: (1) the tendency for the molds to dry on the outside surfaces if left several days before pouring, and (2) the roughness of the casting surface, if the metal is poured too hot. Both of these undesirable properties are eliminated to a large extent after the sand is used several times. After 75 heats, the sand molds better, produces a smoother casting, and shakes out easier than the new mixture of sand.

UTILIZATION OF WASTE STEEL FINES FOR GRAY IRON MOLDING

14. In steel foundries exercising careful sand control, large amounts of fines are regularly removed from the molding sand to maintain the highest possible permeability. The amount of these steel fines, removed as waste material, varies from 200 to 500 lb. per ton of steel castings produced.

15. It has been suggested by H. H. Blosjo, metallurgist, Minneapolis Electric Steel Castings Company, that this waste sand might be useful for making molds for gray iron castings. It is reused to some extent by the steel foundry in core mixtures made from 10 per cent fines and 90 per cent St. Peter sand. This results

in (1) a saving of new sand, (2) a reduction in the amount of steel fines to be hauled away, and (3) the imparting of a certain amount of green strength to the core mixtures so that less cereal binder is used. The presence of steel fines is beneficial in preventing "cat whiskers."

16. The molding sands used at the Minneapolis Electric Steel foundry are made from silica sand, bentonite, cereal binder, and water. The waste steel fines are removed from the used sand in a conditioning unit. Table 5 shows a typical sieve analysis of the steel fines and the sand from which they are removed. Five to 10 tons of the waste material are discarded at this plant each week, corresponding to a loss of 330 lb. of sand per ton of finished steel castings.

Table 5

SIEVE ANALYSIS AND FINENESS NUMBER OF SILICA SAND, STEEL FINES, AND FLOOR SAND

	Per Cent Retained		
Sieve No.	*Silica Sand*	*Steel Fines*	*Floor Sand*
6	0.0	0.0	0.0
12	0.0	0.0	0.0
20	1.1	0.8	0.4
30	14.6	2.9	1.6
40	39.4	10.0	7.5
50	32.1	22.0	7.0
70	11.0	31.6	33.2
100	0.8	21.9	17.5
140	0.3	6.2	8.9
200	0.0	0.6	5.0
270	0.0	0.1	1.0
Pan	0.0	0.1	7.6
A.F.A. Clay	0.7	3.8	10.3
Total	100.0	100.0	100.0
A.F.A. Fineness No.	34.2	53.2	81.1

17. The steel fines have been used for the past year at the University of Minnesota foundry for making molds for gray iron castings weighing up to 200 lb. Part of the bond in the steel fines, as received, is bentonite, as indicated by the swelling of a batch of sand on being moistened. The proportion of bentonite cannot be determined conveniently by any means known to us, although a test method is now under consideration. Rebonded with 2.0 per cent bentonite and 1.0 per cent cereal binder, this

sand is excellent for student use where larger castings are concerned, since it can be rammed hard without dangerously decreasing the permeability. It is also satisfactory for machine molding.

18. When it is required to prepare a sand of definite permeability to be used in an investigation of the permissible variation in this property, blends of floor sand and steel fines are prepared in 150-lb. batches in a 3-ft. muller.

19. The floor sand is the accumulation of all mold-making materials, including a natural bonded molding sand, a small amount of high-clay sand used in sea coal facings, residual sea coal, and a small amount of pitch. Floor sweepings and "burned-on" sand are excluded from the floor sand insofar as it is practical.

20. The blends have reached their present state of development as the result of mixtures prepared in an 18-in. laboratory muller. The sands are dried at 250°F. and cooled. Four thousand grams of sand are mulled 3 min., the proper moisture addition made, and the wet mixture mulled for 5 min. Bentonite additions are made in the initial dry mixture.

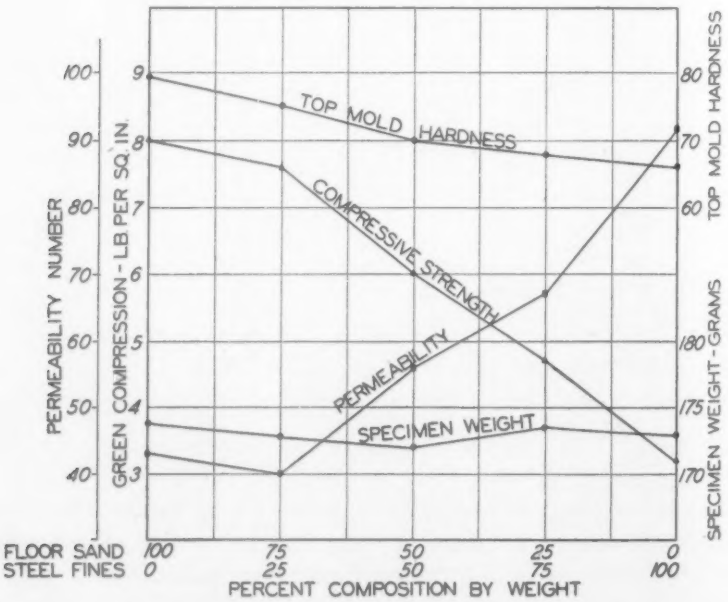

FIG. 8—SOME PROPERTIES OF BLENDS OF FLOOR SAND AND STEEL FINES. MOISTURE, 4.5 PER CENT.

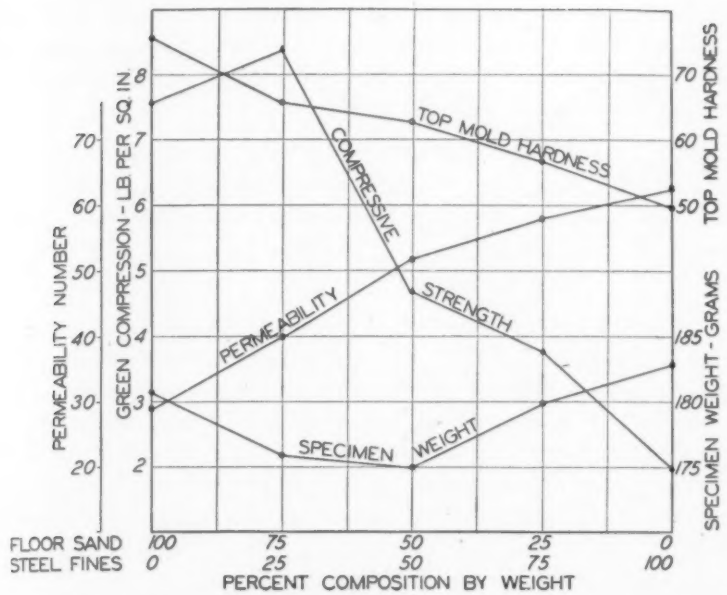

FIG. 4—SOME PROPERTIES OF BLENDS OF FLOOR SAND AND STEEL FINES. MOISTURE, 6 PER CENT.

21. Before the preparation of specimens, the mulled batch is screened through a ¼-in. mesh riddle. Some of the properties of the various mixtures have been determined and are reproduced in part in Figs. 3, 4, 5, and 6. These charts show some of the properties of the two base sands and three blends at two moisture contents.

22. The plotted points are the averages of five values in close agreement. Top mold hardness was obtained with a mold hardness tester, readings being taken in four places 90° apart, midway between the center and the periphery on top of a standard A.F.A. sand specimen.

23. Table 5 shows the sieve analysis and fineness number of the floor sand and the steel fines. A comparison of these analyses with corresponding permeability values on any of the four graphs shows that the permeability of a sand is affected markedly by the presence of the finer grain sizes, non-swelling clays, and an unfavorable grain distribution.

24. As reported previously by Dietert[2], and Briggs and Morey[3], moisture in excess of the amount required to give a de-

sirable temper to a sand, increases the density of the rammed specimen, decreases the permeability, mold hardness and compressive strength. This is evident if corresponding blends are compared in Figs. 3 and 4, and in Figs. 5 and 6.

25. The permeability of a sand may be increased by the proper addition of bentonite, which swells greatly on wetting and forces the sand grains apart, thereby increasing the size of the interstitial tubes. Compare Fig. 3 with Fig. 5, and Fig. 4 with Fig. 6.

26. The optimum temper of a sand is that moisture content which, with the proper mixing treatment, gives a sand of minimum density. On the basis of lowest specimen weight, the sands represented in Fig. 5 should have better properties than the sands of the other three series. Consideration of the graphs shows that this is true. Table 6 summarizes the properties presented in Fig. 5 and shows that these sands, rebonded with small additions of bentonite, and containing 4.5 per cent moisture, are suitable for green sand molding in the gray iron foundry.

RAPID DRYING OF SYNTHETIC SANDS

27. The rapid drying, common to synthetic sands, is observed in the case of the No. 10 mixture of St. Peter sand, and in the

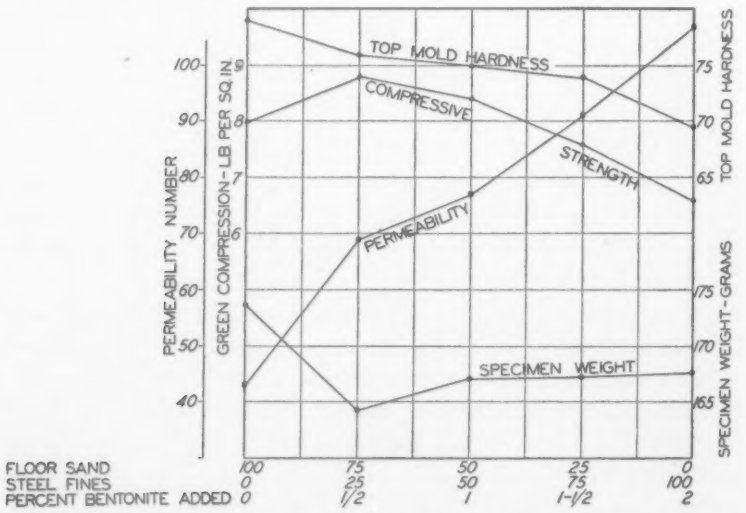

FIG. 5—SOME PROPERTIES OF BLENDS OF FLOOR SAND AND STEEL FINES REBONDED WITH BENTONITE. MOISTURE, 4.5 PER CENT.

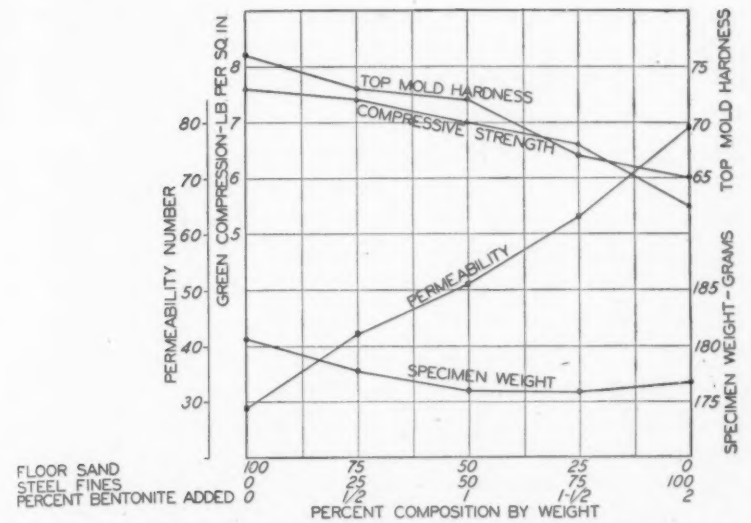

FIG. 6—SOME PROPERTIES OF BLENDS OF FLOOR SAND AND STEEL FINES REBONDED WITH BENTONITE. MOISTURE, 6 PER CENT.

Table 6

SUMMARY OF PROPERTIES SHOWN IN FIG. 5

Per Cent Moisture = 4.5

Floor Sand, per cent	100	75	50	25	0
Steel Fines, per cent	0	25	50	75	100
Bentonite Added, per cent	0	½	1	1½	2
Permeability	43	69	77	91	107
Compressive Strength, lb. per sq. in.	8.0	8.8	8.4	7.6	6.6
Top Mold Hardness	79	76	75	74	69

blends of floor sand and steel fines. It is of interest to note that, as the proportion of steel fines and bentonite increases in the blend, the drying effect becomes greater; that is, the sand becomes less "natural" and more "synthetic." While rapid drying is an important factor in air-dried and oven-dried molds[4], it is not desirable in the case of hand molding requiring patching, careful pattern drawing, and placing of cores.

28. An investigation was recently started to study the rapid drying of synthetic sands. The problem is being attacked from two view points, (1) factors influencing drying and (2) means of preventing drying. Glycerine is reported[5] to be useful in preventing drying, and our preliminary work substantiates this. Experi-

ments dealing with the effects of a wetting agent, asbestos fibre, alkali and alkaline earth oxides, hydrous silica and hydrous ferric oxide are also in progress.

29. The wetting agent is of special interest because of its penetrating and emulsifying properties resulting from its effect in decreasing the surface tension of water, which, in turn, is related to its chemical composition. Important effects on the colloidal properties of clays were expected since the hardening of starch pastes has been materially retarded by small percentages of wetting agents.

30. Work finished to date indicates that, while the wetting agent is not effective in reducing drying, it has been found in several instances that, at a critical moisture content, the green compressive strength increases to double the strength obtained with plain water and the same percentage of bond.

CONCLUSIONS

31. Although this paper is considered only a progress report, several conclusions can be drawn.

1. The St. Peter sandstone formation offers to the mid-western states an inexhaustible supply of nearly pure silica sand.

2. This sand can be used as a base for synthetic molding sand as well as for core sand.

3. Fines removed from molding sands of high permeability can be used as molding sands of lower permeability.

4. These fines can be used as blending sands to control the properties of molding sands of similar composition.

ACKNOWLEDGEMENTS

32. The authors wish to acknowledge the assistance and co-operation of Prof. George A. Thiel, geology department, in supplying geological information, and the work of William Brennan, Joseph Pagani and William Mackay, students in the Institute of Technology, University of Minnesota, in conducting many of the tests. They also wish to thank the Geological Society of America for permission to use the map shown in this paper.

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DISCUSSION

Presiding: D. L. PARKER, General Electric Co., West Lynn, Mass.

MR. HOLTBY: This paper can be considered as a progress report, as we have done considerable work since this paper was written. I might say that this is about one-third of what we expect our publication on this subject to be.

N. J. DUNBECK[1] *(Written Discussion):* I have read with much interest this paper covering experiments with synthetic molding sands. The author's experience that wetting agents are not very effective in reducing the drying out of synthetic sands confirms the results obtained in our laboratory on similar work which included a considerable number of such agents. We did not, however, note any sharp increase in green strength as mentioned in the paper. We would be much interested in knowing what agent gave this result and the proportions which were found effective.

In further work on the prevention of drying out, I suggest trying ordinary fuel oil which is the material most widely used for the purpose in actual practice. One investigator also reported very good results from the use of calcium chloride.

MR. HOLTBY and MR. SCOBIE *(Written Discussion):* We have attempted to produce a series of clay blends with the idea of attaining any desired combination of green and dry strength. In this work, a good grade of bentonite was mixed with a "plastic clay" used in brick manufacture. The results are not immediately available but it is recalled that the "plastic clay" contributed little to the strength of the mixture. Hot strength tests were not conducted.

[1]Vice President, Eastern Clay Products Co., Eifort, O.

The wetting agent which has been used is Aerosol OT Aqueous 10 per cent. Moisture is added as a solution of the 10 per cent Aerosol, the solution being prepared in the proportion 1:99 by weight. The work will include a solution 5:95 by weight. Only bentonite has been used as a bond.

N. J. DUNBECK: I think everyone who has used synthetic sands has noticed the fact that they do have a tendency to dry out somewhat more rapidly than natural sands, or at least the loss in moisture and the drying out is more apparent. We normally use a lower percentage of tempering water in our synthetic sands and when we lose a certain amount of that water in hot weather, it is apparent; we notice it.

The one item of particular interest to me in the paper was that the use of wetting agents in some cases was not only effective for its purpose but also resulted in higher strength.

MR. HOLTBY: The property of this wetting compound to increase the strength was noted in several cases and immediately it raised the question as to whether it was true. We re-ran the test and I have a few test results here that I would like to read. In this particular test, the mixtures were bonded with 4 per cent dried bentonite and were prepared from the St. Peter sandstone. After mulling dry for 3 min., the moisture was added and mulling continued for 5 more min. exactly as we had tried before. The sand was riddled before preparation of the specimens. This wetting compound that we are using was added.

The graph in Fig. 7 shows a jump from an expected pre-strength of about 7.5 lb. per sq. in. up to as high as 13.5 lb. per sq. in. at a moisture content of 2 per cent using this wetting compound solution, which is a 0.1 per cent Aerosol solution.

MEMBER: Did the authors try any other wetting agents, and a second question is, did they find that the addition of the wetting agent changed the moisture content at which they got this optimum green strength? In other words, did they have to use more or less moisture in order to develop the highest green strength when they used the wetting agent?

MR. HOLTBY: The answer to the first question is no. This was the only wetting agent that we have had in stock and is part of the test that is yet to come. We did not have much time to prepare this after the paper was originally written. We expect, however, to test some of the other wetting agents.

As to the answer to the second question, I might explain this curve in Fig. 7 in more detail. Without the wetting agent at 1 per cent moisture, with the 4 per cent bentonite mixture, the compressive strength started at 8 lb. per sq. in. and dropped off gradually to about 3.5 lb. per sq. in. at 8 per cent moisture. That was a gradual drop-off. Now, with this same material, but adding this wetting agent with the water as a

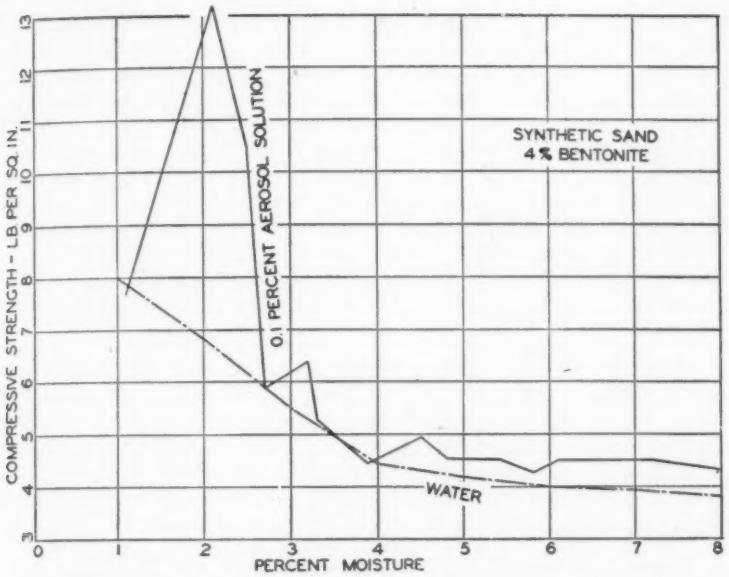

FIG. 7—EFFECT OF WETTING COMPOUND ON THE STRENGTH OF SYNTHETIC SAND.

one-tenth of one per cent solution, we started out with 8 lb. per sq. in. at the 1 per cent moisture content and then at 2 or little bit better than 2 per cent moisture, it went way up to a little over 13 lb. per sq. in. Then, at about 2.5 or 3 per cent moisture, it drops back down and comes very close to the curve where the wetting agent was not used. In other words, the two curves lay close together except at the one jump point, which goes way up to a high peak. This test has been repeated a number of times.

MEMBER: Have the authors used any of this sand without any cereal binder and, if so, with what results?

MR. HOLTBY: Yes, we used some of this sand without cereal binder, but I do not have that data with me. The cereal binder was added in our mix to try to keep it from drying out. In other words, we find with the cereal binder, using a bentonite mixture, that our sand does stay in a workable condition. That is, you can patch it, the edges do not dry out and after the mixture leaves the mixing machine or lays on the floor, it remains moist for a much longer period of time, at least apparent moisture to the workmen.

CHAIRMAN PARKER: In regard to the roughness of the castings mentioned by the authors, I was wondering if they, after a series of heats, had made any attempt to study the appearance of the grains. In other words, was there a conglomerate grain present? Did it seem to build up?

MR. HOLTBY: I understand that question to mean the grain of sand. Were these binders still on it? We did not look at that under the microscope very carefully. We checked the grains without the binder. We seem to feel that after the sand has been used, it has somewhat better properties. In other words, it works better, it is not sticky like the new mix. It seems to flow better in the mold, if you want to use that term. In other words, in the molding machines, with a new batch of material just mixed up, our No. 10 mix, and with a deep cope or deep drag, we had some trouble alongside a steep pattern to get the surface of the casting smooth. After use, this material seemed to settle down in the mold under normal jolting or squeezing and produced a much superior surface on the casting. But I have not noted any conglomeration of the grains with the binder.

H. F. SCOBIE *(Author's Written Closure):* Since this paper was written, about 20 heats have been poured in the same No. 10 mix described above. The same satisfactory behavior has been noted. Periodic additions of binder have been found to be better practice than rebonding before each heat. Castings of all shapes and sizes from a fraction of a pound to 150 lb. and with sections from ⅛-in. to 5-in. have been produced with good results.

Microscopic examination of the sand after almost 100 heats reveals no change in grain shape due to thermal or mechanical shock. The new sand has a generally rounded shape, although some of the grains are egg or pear-shaped. The grains are quite smooth and not stained by limonite. After use, the grains are the same general shape, discolored, some are slightly fused, and there is a tendency for an occasional small grain to stick to a larger one. There is a great disparity in size between these satellite grains and the large grains to which they are attached. There is no agglomeration of like sized particles nor more than two or three satellite grains on a grain of larger size.

Chaplets and the Steel Casting†

HOWARD F. TAYLOR* AND EDWARD A. ROMINSKI,* ANACOSTIA, D. C.

The influence of chaplets on the soundness and integrity of steel castings has been a point of grave concern since their advent with the industry. The present study is an initial step in the interests of bringing to light some of the many factors operative in governing the fusion characteristics of the various types of chaplets available. Hard and fast rules cannot be established on the basis of work done to date, but many generally-ignored principles of chaplet handling are considered in the hope that more attention and care will be given their application in the future than has been the case in the past. Many castings are rejected, or repaired, daily because of lack of fusion or some other defect occurring in the vicinity of a chaplet or internal chill. It should be borne in mind that the former is really only a specialized form of the latter. The widespread antipathy toward the use of the internal chill is well founded, but it is highly possible that basic research would, in time, obviate this. A little care in the selection and use of chaplets will go a long way toward their better performance, even though the ideal one to fit all conditions has yet to be discovered. The increased demands for and on pressure castings dictate that the maximum attention must be given every detail of their manufacture. Hence, the present contribution, which suggests a suitable test piece for use in the individual foundry, outlines certain considerations and principles, and presents the results of a number of specific tests offering experimental evidence of some of the defects which may result in steel castings through improper selection and/or lack of care in the use of chaplets. Nickel, copper, silver, cadmium, and combination silver-cadmium plated chaplets were tested as well as aluminum sprayed, aluminum dipped, calorized, tin dipped and some of the new silicon impregnated variety. A theory for the mechanism of fusion of low carbon steel chaplets is presented.

* Division of Physical Metallurgy, Naval Research Laboratory.
† Published by permission of the Navy Department.

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GENERAL CONSIDERATIONS

1. A great many chaplets of a variety of sizes and designs are used regularly wherever castings are made which necessitate internally supported cores. A chaplet, ordinarily is, a spool-shaped piece used in a mold for supporting or spacing cores. Some examples are shown in Fig. 1. Many castings are rejected daily because of defects occurring immediately adjacent to, or in the near vicinity of these chaplets, either because of lack of fusion at the chaplet-cast metal interface or as a result of porosity caused by any of several factors. Such defects become especially deleterious if the casting is to be used in some type of pressure installation, as in the case of valve bodies. Any defective areas must be chipped or ground out and replaced by weld metal, necessitating considerable expense and loss of time. The indication is clear that too little attention ordinarily is given to the many factors which influence the successful application of chaplets. Altogether too often a stock supply is supplied, upon which every molder draws, regardless of any consideration other than the initial necessity of supporting a core.

Factors Affecting Chaplet Reactions

2. Some of the factors affecting chaplet reactions may be listed as follows:

(a) Diameter of chaplet supporting rod.

(b) Temperature of the cast metal surrounding the chaplet.

(c) Design of chaplet, especially the supporting rod.

(d) Material of the chaplet.

(e) Composition and quality of plated, dipped, or alloyed exterior chaplet covering.

(f) Cleanliness of surface both before and after coating.

(g) Mold conditions.

Effect of Chaplet Volume on Fusion

3. It is common among practical foundrymen to regard strength and rigidity as chief criteria in choosing chaplet material for a certain job. As a result, many types are employed daily which are unnecessarily large for the weight of core supported and which are too large to be properly fused with the cast metal at normal pouring temperatures. The ratio of cold chaplet volume to the

available heat content of the molten metal surrounding it, is often so large as to preclude any possibility of fusion. This would indicate that a certain rather critical chaplet diameter might exist for a particular metal temperature, but, for reasons disclosed later, it appears that a fairly large latitude prevails. It would be illusory to attempt to prescribe specific metal temperatures for any particular application, since control is not normally practiced within very close limits.

4. However, many castings have been noticed in which not only were chaplets unfused, but large open gaps were present for a radius of two or three inches around the chaplet stem. This could only result from too low pouring temperatures or to the metal having to run too far in the mold. With present improvements in fluidity tests and optical and photoelectric pyrometers, this extreme condition should be obviated conveniently. Thin walled castings are the chief offenders in this regard, and, from the standpoint of proper fusion the mass of metal surrounding the chaplet, must be considered as well as metal temperatures. When it is known that metal temperatures will be unavoidably low, some type of chaplet should be chosen which has a lower fusion point than those normally provided. Such a type is the silicon impregnated variety to be discussed later.

5. It often happens that chaplets are chosen with a stem too small for the weight of core supported and the temperature of metal poured. This causes premature melting and complete fusion to the extent that the core shifts before the metal has solidified enough to support it. This occurs with less frequency than the selection of an oversize type because of the desire of a molder to be on the safer side, knowing that unfused chaplets or porosity caused by them are repairable by welding, whereas a shifted core automatically destroys further use of the casting. However, this is a point worth bearing in mind. Certain low melting point chaplets to be put on the market might give trouble if used with the same judgment as plain low carbon steel types. Core supporting properties, from the standpoint of rigidity alone, might be equal in each case, or even better in the alloyed chaplets, but the lowered melting point of the one would cause trouble unless used with care.

Effect of Shape

6. There seems to be a definite trend toward the making of chaplet stems into some form of corrugated, rather than plain,

surfaces. No doubt this is done with the expectation that the roughened, or irregular, surfaces are an aid to fusion, and possibly to a certain extent as being more appealing to a buyer. It is logical to assume that sharp points and patterns in relief may fuse more readily than a plain surface, but experimental evidence to be presented does not always bear this out. What advantage might be gained in promoted fusion may be nullified by the formation of gas pockets in the sharp corners and depressed sections.

Effect of Material

7. The material from which the chaplet is made should be a point for consideration. Chaplets are ordinarily made of low carbon steel with the stems stamped, rolled, or machined into the desired shape. The end pieces are stamped, placed on the stem, and pressed into a tight fit to make a rigid assembly. The relatively high melting point of the low carbon steel prevents too rapid melting and is cheap, conveniently shaped, and easily made.

8. Some high alloy chaplets are on the market in which fusion results in an alloying action in the cast metal for a considerable distance around the chaplet-cast metal interface. This would affect the properties of this region according to the nature of the element and might be advantageous in some cases and bad in others.

Effect of Surface Preparation

9. In regard to special preparation of chaplet surfaces, many different methods have been tried and some are listed and discussed herein. Special treatment is necessary for a variety of reasons. If the plain, low-carbon steel chaplets were left exposed to the air without any protective covering, they would rust and accumulate moisture. Upon use in a casting, excessive porosity would result at the fusion zone and for considerable distances into the body of the metal. This would be caused by the reaction between the carbon of the steel and the iron oxide of the scale, with the formation of carbon monoxide. Also any moisture present as such in the scale, or on the surface of the chaplet, would cause defects and retard satisfactory fusion. The surfaces of uncoated chaplets could be cleaned in acid or sand blasted immediately prior to use, but this would be both inconvenient and

subject to lack of care. To obviate this, chaplets are coated and prepared in several ways.

Coatings

10. Electroplated surfaces include nickel, copper and cadmium in thicknesses from 0.0005 to 0.01-in. In a series of tests conducted by cooperating industrial foundries, the copper plate was found to give the most consistently good results of any tried. Included in the work was a tin dipped series, but none of the silicon impregnated variety. What type or how extensive the investigation was is not known.

11. So-called "dipped" chaplets are prepared by dipping the cold pieces into a bath of molten metal, such as tin, and allowing a thin layer to adhere, after which they are removed and cooled. Opinion varies greatly among foundrymen as to the relative merits of this type.

12. Silicon impregnated chaplets can be prepared by the process developed by Dr. Harry Ihrig, the chief function of the process being to cause silicon to alloy with the iron by diffusion at some temperature below the melting point of either one. So far as known, impregnation of chaplets has only been tried with silicon, and these are of very recent origin and are not yet marketed commercially.

13. Dr. Ihrig's process of silicon impregnation consists of heating iron or steel to 1700-1850°F. in the presence of silicon carbide and chlorine. Forged, rolled, or cast low-sulphur, low-carbon steel is preferable for the process. Ferro-silicon can be satisfactorily substituted for all or part of the silicon carbide. The material used in this work was made by this process. Ordinary carburizing equipment can be used and chlorine is added after the parts are at temperature. Rotary or pot type furnaces are equally applicable. The carbide should be in contact with the metal and, while the mechanism of the process is not clearly understood, it is believed that nascent silicon is liberated and diffuses readily into the metal. Cases from 0.005 to 0.1-in can be obtained. The silicon-rich iron case, containing as much as 14 per cent silicon, has a relatively low melting point and would fuse, or perhaps completely melt, very readily. Such being the case, fusion would result at a specific temperature with considerably larger impregnated chaplets than plated or dipped types.

Cleanliness

14. Cleanliness of chaplet surface, both before and after coating, merits much greater consideration than it is commonly given. Any scale or dirt present prior to coating will be covered up by a thin layer of deposited metal and is a potential source of porosity in the casting. Also a very thin or loose layer, or at worst none at all, might adhere in the vicinity of the scale, allowing more oxide to collect. In impregnated varieties, this is not so important and the corrosion resistance of this type is excellent.

15. Rust is only one of many forms of uncleanliness that may lead to poor chaplet fusion. Any oil or moisture present would cause porosity, the former being spilled accidentally or rubbed from the hands of any person handling the material. Moisture could result from water spilled over the chaplet or by condensation from the atmosphere before or after being put into the mold. Fine dust often collects in irregularities or even on smooth surfaces, if chaplets are stored in open containers on the foundry floor, and may account for some porosity and much of the slag sometimes noticed in unfused specimens.

Effect of Molding Conditions

16. Conditions of molding have an important influence on the satisfactory functioning of chaplets. If a green sand mold containing chaplets is allowed to stand for a very long interval before pouring, moisture from the humid atmosphere of the mold condenses on the cold metallic surface and either remains as water or promotes rusting, each capable of causing faulty castings. In cases where green sand molds are oven dried with cores in place, very favorable conditions for rusting are presented, and any spot of porous or discontinuous coating would be readily attacked. The flaky nature of the silicon impregnated chaplets tested, present good possibilities for the absorption of moisture and the collection of dust.

17. The work is presented in two parts, since it was initially started with the intention of checking only those types of chaplets in general use. Part 1, therefore, includes that phase of the investigation. Not knowing the exact history of the pieces and the details of their preparation, and desiring several types not commercially available, it was decided to further study the behavior of chaplets prepared under carefully controlled laboratory conditions. The results of this work comprises Part 2.

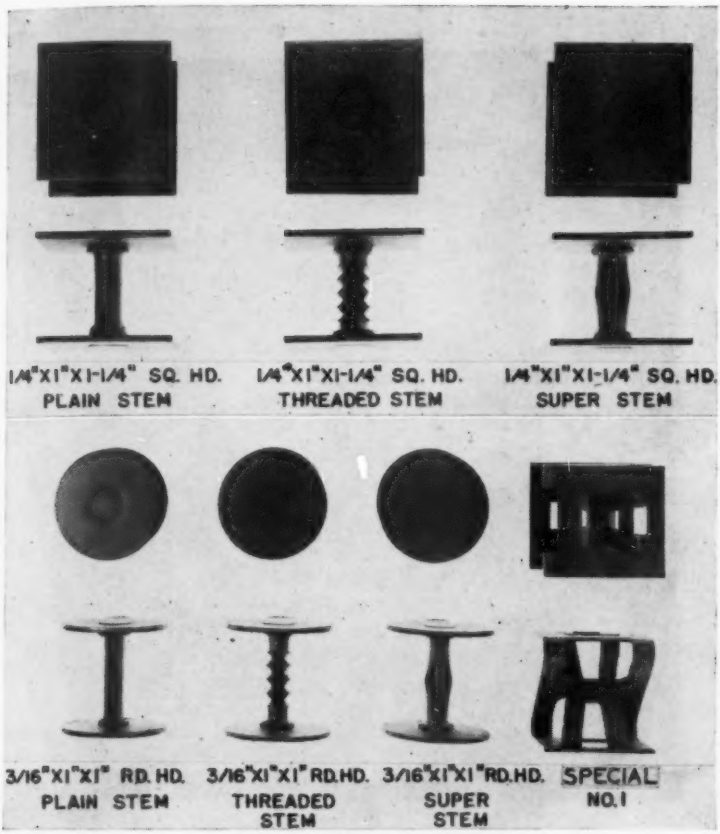

1/4"X1"X1-1/4" SQ. HD. PLAIN STEM
1/4"X1"X1-1/4" SQ. HD. THREADED STEM
1/4"X1"X1-1/4" SQ. HD. SUPER STEM
3/16"X1"X1" RD. HD. PLAIN STEM
3/16"X1"X1" RD.HD. THREADED STEM
3/16"X1"X1"RD.HD. SUPER STEM
SPECIAL NO.1

FIG. 1—CHAPLET STYLES.

Part I — Tests on Commercial Chaplets

EXPERIMENTAL PROCEDURE

18. For this part of the investigation, four different styles of chaplets were obtained from a reputable manufacturer. These are shown in Fig. 1. The plain stem styles were only obtained for the second part of the investigation and the Special No. 1 type (Fig. 1) was not received in time for testing. This is included only as a matter of interest in design.

19. Copper plated, tin dipped and silicon impregnated types were obtained in the 1/4 and $\frac{3}{16}$-in. coarse thread and superstem designs and tested as received from the factory.

Molds Used

20. The 24 pieces were placed in green sand molds without supporting any cores, and held in place by the pressure of the cope bearing on the flat upper plate of the chaplet. It was not thought necessary to use a core since the main aim of the investigation was the adaptability of the three different series from the standpoint of satisfactory fusion. Fig. 5 illustrates the casting design and the position of the chaplets in the casting. The design of the test piece was considered to furnish uniform conditions for each specimen tested.

21. The molds were made of a synthetic sand mix consisting of washed and graded silica sand, bentonite and water. They were prepared approximately 3 hr. before pouring and left open for about 2½ hr.

Furnace and Pouring Practice

22. The six molds containing the 24 chaplets were placed in order on a long truck of the small gauge railroad type which could

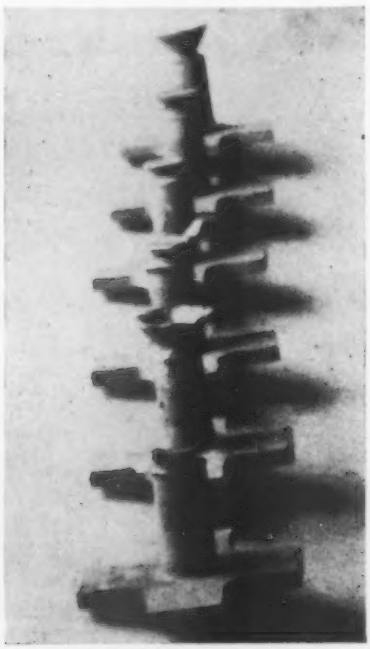
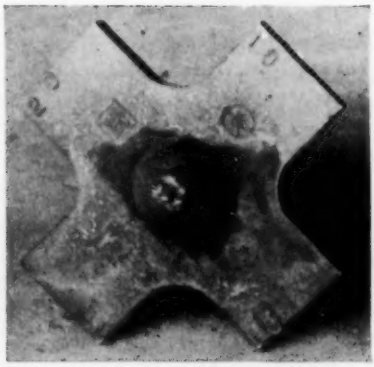


FIG. 2—TEST PIECES AS CAST.



LEFT—FIG. 3—COPE SIDE OF TEST PIECE. RIGHT—FIG. 4—DRAG SIDE OF TEST PIECE.

be pushed along in front of the melting unit for more rapid pouring of the successive castings. The furnace used was an induction, tilting-type, of 300-lb. capacity and lined with a fused silica crucible. The metal was an average analysis of 0.21 per cent carbon, 0.78 per cent manganese, 0.32 per cent silicon, and 0.45 per cent nickel. As soon as the metal was melted, the final addition of ferromanganese and ferrosilicon was made and the temperature raised to 2925°F. The temperature was measured with an optical pyrometer, calibrated against a thermocouple in previous baths of steel. At this point, the power input was decreased slightly to hold the steel at this temperature while each of the six molds were poured in rapid succession.

23. Fig. 4 shows the castings after being shaken from the molds and indicates that each was poured to the same height in the riser. Figs. 3 and 4, show an individual casting from the cope and drag side respectively. The round catch basin was placed directly under the pouring gate to allow a reservoir of metal to build up and equalize the flow into the four arms. The positions of the chaplets are clearly seen, as well as the method of marking their identity.

Specimens

24. After cooling, the castings were sawed into specimens 1½-in. square, around the region of the imbedded chaplet, as shown in Fig. 6. This piece was then cut in half to give a section across the chaplet stem and the second half cut lengthwise through the rod. The resulting specimens are as shown in Figs. 7 to 12

inclusive. The cut surfaces of the specimens were ground flat and finish polished on coarse metallographic emery paper. This was considered satisfactory to give the general information desired in the initial investigation. For the more precise metallographic analy-

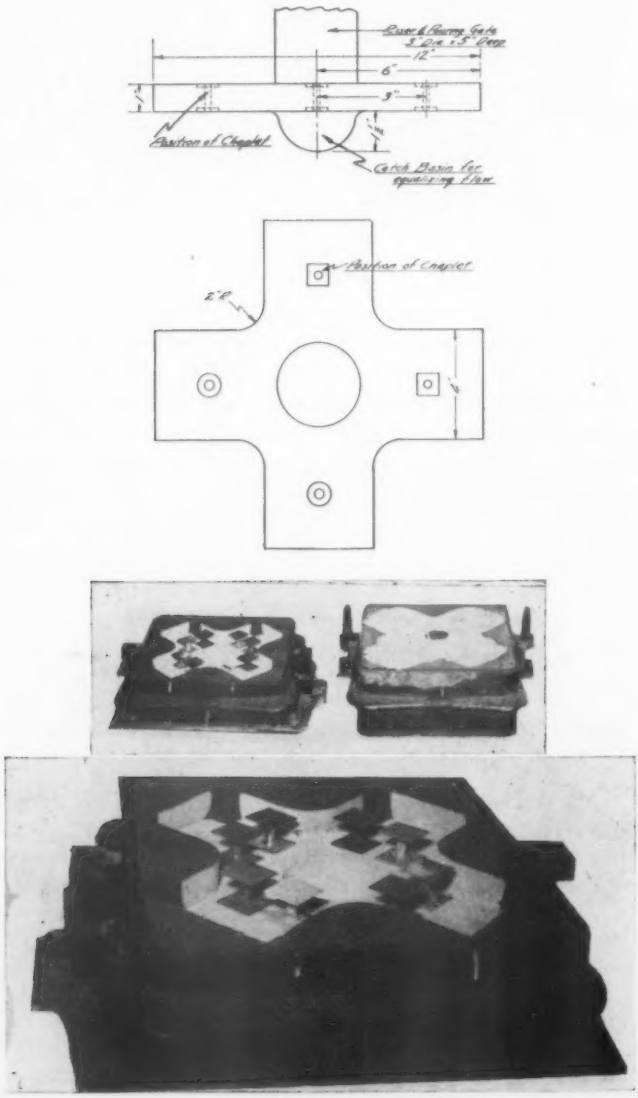


FIG. 5—TOP—DETAILS OF TEST CASTING FOR CHAPLET STUDY. BOTTOM—WASHED AND DRIED TEST MOLD WITH CHAPLETS IN PLACE.

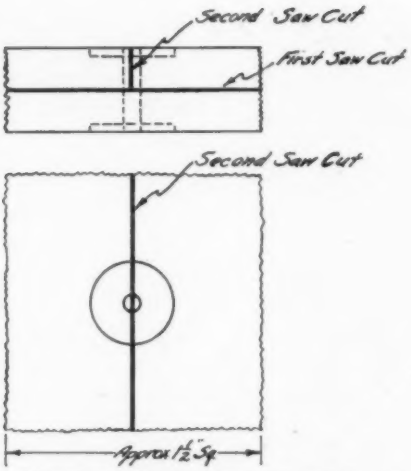


FIG. 6—METHOD OF CUTTING SPECIMENS FOR EXAMINATION.

sis, discussed later, considerably more care was taken in preparation. For the etched specimens, a 3 per cent solution of nitric acid in ethyl alcohol was used. A hardness survey was made across the face of one of the pieces in which a silicon impregnated chaplet had been used. This was done with a Vickers-Brinell machine using a diamond pyramid.

DISCUSSION OF RESULTS

25. The macrographic analysis of work done to date is presented in Figs. 7 to 12, inclusive. The duplicate series of Figs. 7 and 8 were made to check the ability to reproduce results in subsequent castings not poured in order; that is, placed out of order on the pouring truck. Figs. 9 and 10 present a duplicate set cast in order, one after the other, while Figs. 11 and 12 include the hot tin dipped and silicon impregnated series.

26. The photographs of the unetched specimens give a much clearer indication of the degree of fusion in each case than do the etched pieces. What appears to be a clear line of demarcation between chaplet material and the cast steel is often only a film of impurity deposited at the interface during etching and accentuated in photographing. A micro-examination revealed this to be the case.

27. In Figs. 7 to 10, it is evident that the degree of fusion desired has not been obtained. Such porosity and lack of fusion as indicated therein might enable a casting to pass an initial test, but

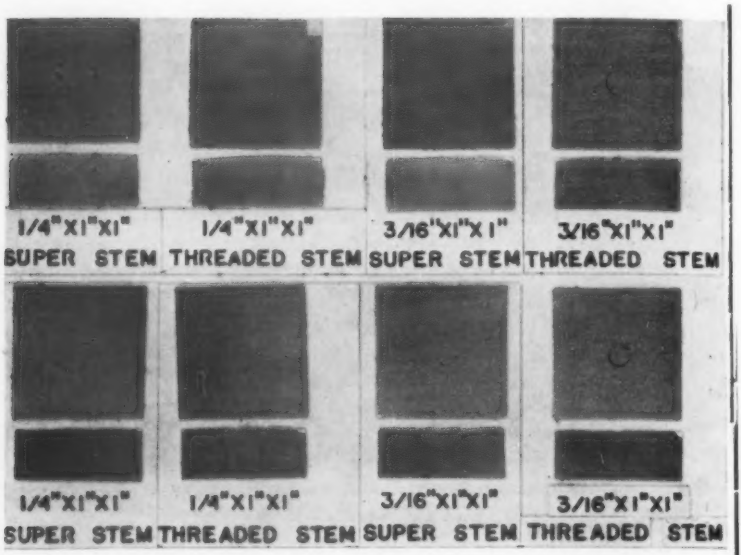


FIG. 7—POLISHED SPECIMENS—COPPER-PLATED CHAPLET SERIES—CHAPLETS COMMERCIALLY BARREL-PLATED—PLATE 0.0005-IN. THICK.

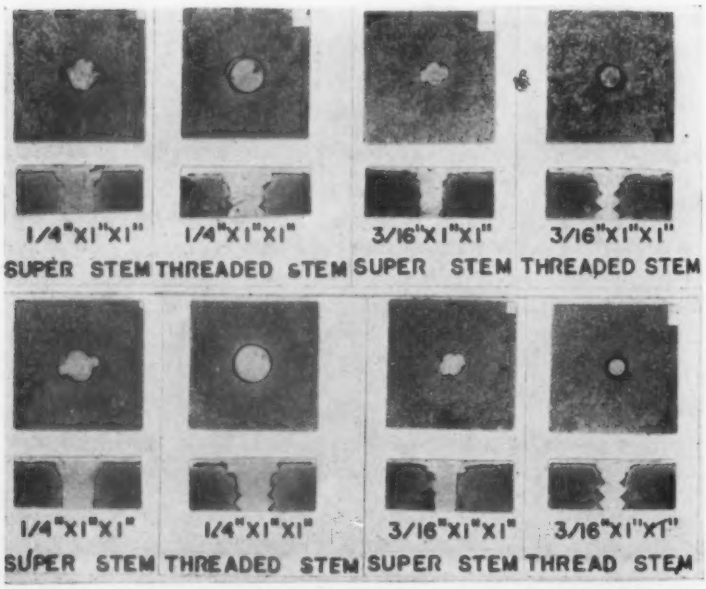


FIG. 8—SAME SERIES AS FIG. 7 BUT ETCHED WITH 3 PER CENT NITAL.

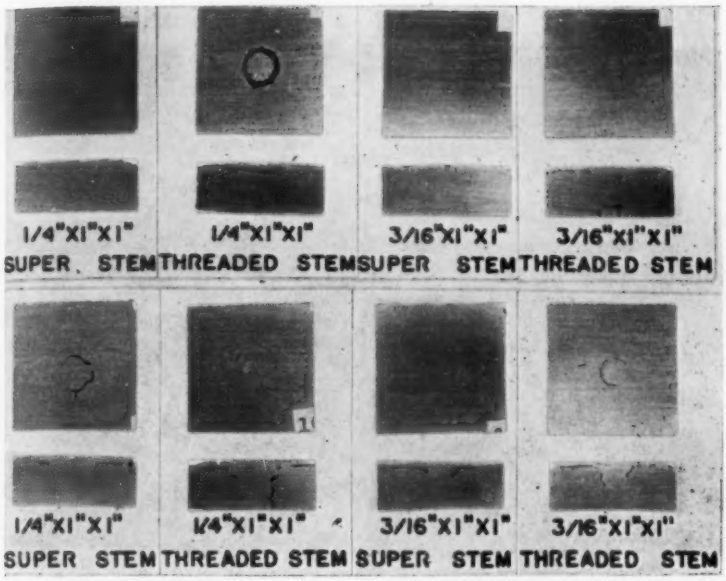


FIG. 9—SAME TYPE CHAPLETS AS IN FIGS. 7 AND 8 WITH PLATE THICKNESS OF 0.01-IN. POLISHED.

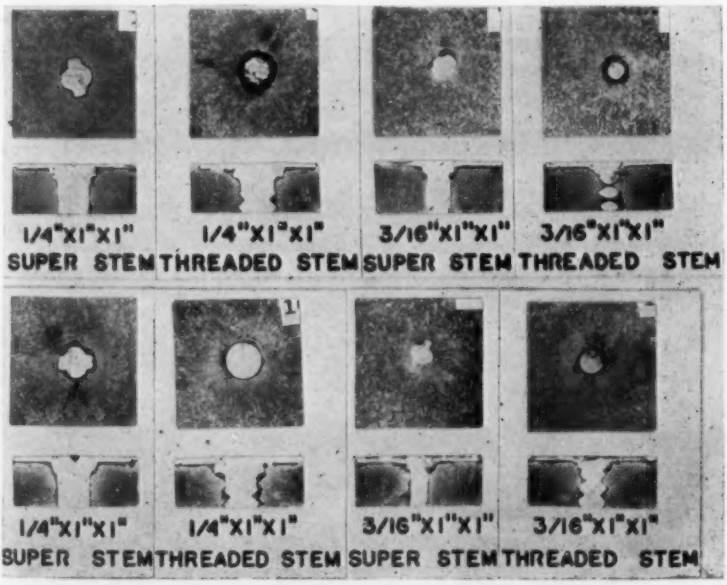


FIG. 10—SAME SERIES AS FIG. 9 BUT ETCHED WITH 3 PER CENT NITAL.

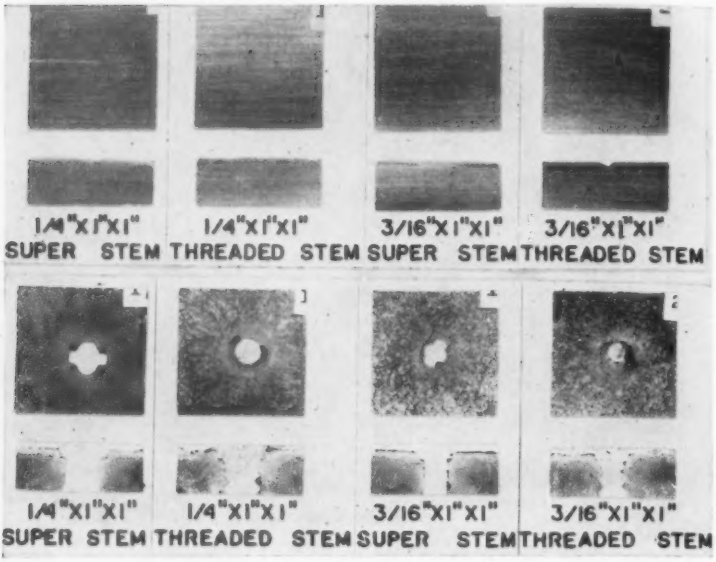


FIG. 11—TIN-DIPPED SERIES. TOP—POLISHED. BOTTOM—ETCHED WITH 3 PER CENT NITAL.

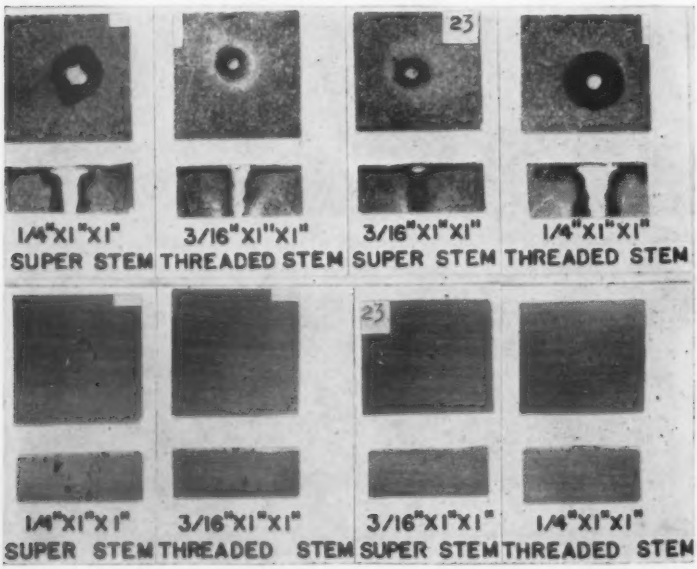


FIG. 12—SILICON-IMPREGNATED SERIES. TOP—ETCHED. BOTTOM—POLISHED.

if the piece is destined for service under steam or hydraulic pressure, failure would ultimately result in many of the cases. Porosity is evident both at the fusion zone and for some distance into the metal.

28. How far away such spots occurred from the region of the chaplet was not determined in the present investigation, but radiographs of many castings indicate that they can be very widespread. This is undoubtedly the result of any dust, oxide, or volatile substance on the chaplet surface which can wash into the metal stream. In some cases, localized fusion has occurred in portions of the chaplet while others show no fusion whatever.

29. The sections showing the cut taken longitudinally through the chaplet rod is placed directly above the corresponding transverse section. From a comparison of each, it is clear that by using only the latter a mistaken conclusion could be drawn. If the cut was made presenting the sharp points of a thread, it would appear that fair fusion was obtained. However, if the cut showed the bottom of the "V," an entirely different interpretation would result.

30. It does not appear from Figs. 7 to 10, that threaded stems give any greater assurance of fusion than the superstem type. It is felt that a plain type with a smooth fillet at the juncture of the stem and plate might be superior to any of the present types.

31. From the present work, little can be said regarding chaplet stem diameter. In certain specimens, differing only in size, about the same degree of fusion is evident in each. In some specimens, the larger even showed the better fusion, indicating that the latitude of chaplet size is wide enough for practical purposes.

32. Figs. 7 to 10, showing copper plated series differing only in thicknesses of coating, indicate no constant difference in fusion. The tin dipped series of Fig. 11 show very good fusion and little porosity. The ¼-in. superstem is a nearly perfect example of the type of union desired. Any of the four specimens of this series would probably function properly under pressure.

33. The silicon impregnated chaplets of Fig. 12 show satisfactory fusion but also a slight amount of porosity, especially in the ¼-in. superstem specimen. These would undoubtedly give satisfactory service from the standpoint of leakage due to porosity. The dark areas around the chaplet, shown in Fig. 12, is

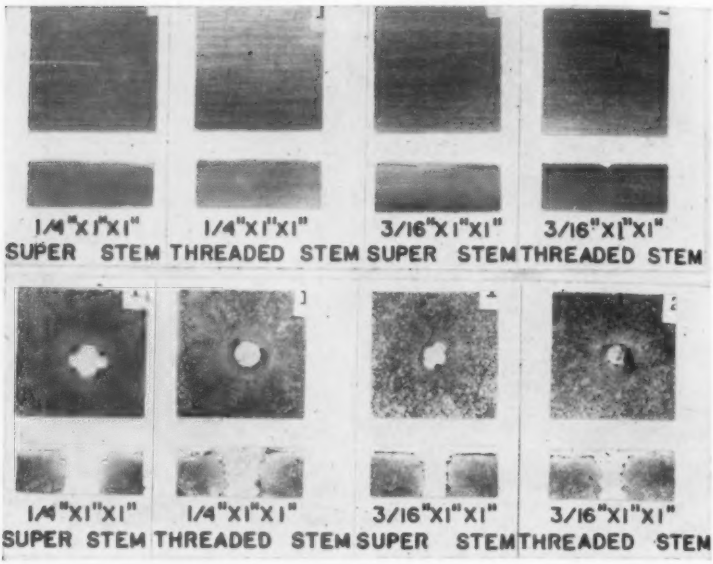


FIG. 11—TIN-DIPPED SERIES. TOP—POLISHED. BOTTOM—ETCHED WITH 3 PER CENT NITAL.

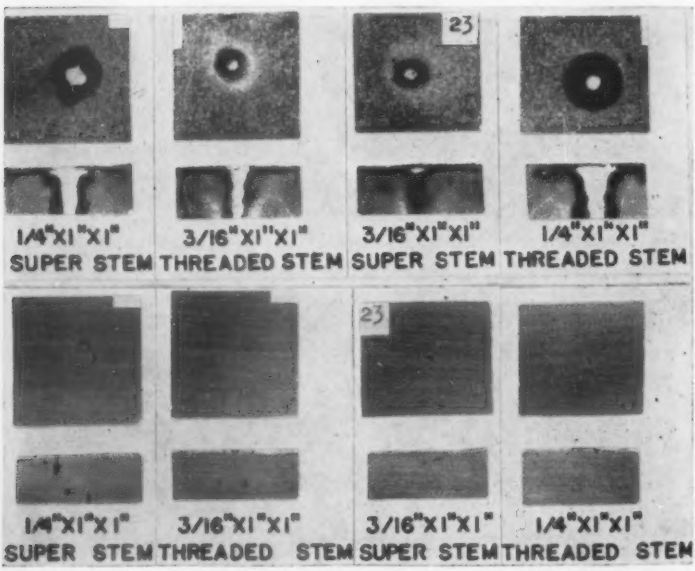


FIG. 12—SILICON-IMPREGNATED SERIES. TOP—ETCHED. BOTTOM—POLISHED.

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30. It does not appear from Figs. 7 to 10, that threaded stems give any greater assurance of fusion than the superstem type. It is felt that a plain type with a smooth fillet at the juncture of the stem and plate might be superior to any of the present types.

31. From the present work, little can be said regarding chaplet stem diameter. In certain specimens, differing only in size, about the same degree of fusion is evident in each. In some specimens, the larger even showed the better fusion, indicating that the latitude of chaplet size is wide enough for practical purposes.

32. Figs. 7 to 10, showing copper plated series differing only in thicknesses of coating, indicate no constant difference in fusion. The tin dipped series of Fig. 11 show very good fusion and little porosity. The ¼-in. superstem is a nearly perfect example of the type of union desired. Any of the four specimens of this series would probably function properly under pressure.

33. The silicon impregnated chaplets of Fig. 12 show satisfactory fusion but also a slight amount of porosity, especially in the ¼-in. superstem specimen. These would undoubtedly give satisfactory service from the standpoint of leakage due to porosity. The dark areas around the chaplet, shown in Fig. 12, is

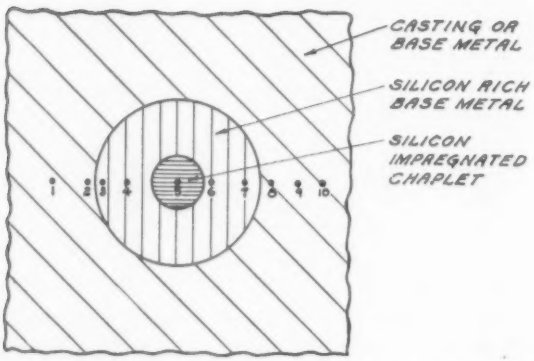


FIG. 13—POINTS AT WHICH HARDNESS READINGS WERE TAKEN AS RECORDED IN TABLE 1.

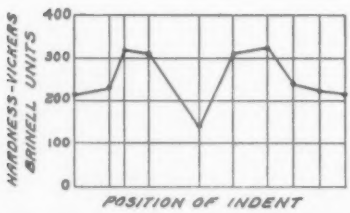


FIG. 14—CHART SHOWING POSITION AND HARDNESS READINGS TAKEN AS SHOWN IN FIG. 13.

base metal into which the silicon from the chaplet has diffused. It might be thought that these regions would be brittle and the casting might crack under sudden, localized shock. In the majority of installations, danger from this source is probably slight, since the usually more ductile material surrounding these areas would absorb most of the shock in isolated cases where it might occur.

34. No actual measurement of brittleness was made but a hardness survey was taken across the ¼-in. coarse-thread specimen in the manner indicated in Fig. 13. The increase in hardness can be taken as a rough criterion, together with the well known embrittling action of silicon. The results of the hardness measurements are given in Table 1 and are plotted in Fig. 14.

35. In forming conclusions from Fig. 12, it should be noted that the longitudinal cut of the $\frac{3}{16}$-in. superstem specimen does not show any chaplet material. This was removed by the saw cut as so little was left unfused.

Table 1

HARDNESS SURVEY ON INFLUENCE OF SILICON-IMPREGNATED CHAPLETS

Position of Indent as Shown in Fig. 13	*Hardness in Vickers Brinell units*
1	215
2	228
3	311
4	309
5	145
6	302
7	318
8	238
9	226
10	219

THE MECHANISM OF FUSION

36. It is commonly believed that fusion between chaplet and cast metal is a case of simple melting, enough superheat existing in the melt to raise the temperature of the chaplet above its melting point. This is a fair belief, since most steel is poured at relatively high temperatures. However, cases are known where cast iron has fused soundly with low carbon steel chaplets, the temperature never reaching the fusion point of the latter. This leads to the necessity for considering some other phenomenon than that of simple melting.

37. Considering the case of cast iron and turning to the iron-carbon diagram, it is obvious that chaplet metal containing 0.10 per cent carbon and no other alloying elements, would have a melting point higher than that normally reached by the cast iron poured around it. In order for fusion to result, some means must be provided for a lowering of the fusion point. When immersed in molten cast iron, it is known that chaplet material will absorb carbon by diffusion and the carbon content of the surface will rise progressively to a high figure. This carburization will go on until the melting point of the chaplet surface has been lowered to correspond to the temperature of the surrounding cast metal. It is obvious that unless some such mechanism exists, fusion in the above instance would be impossible unless the cast iron were held at the fusion point of iron containing 0.05 to 0.15 per cent carbon, *i.e.*, 1490-1500°C. (2714-2732°F.), a temperature too high for the production of sound iron castings. It is known

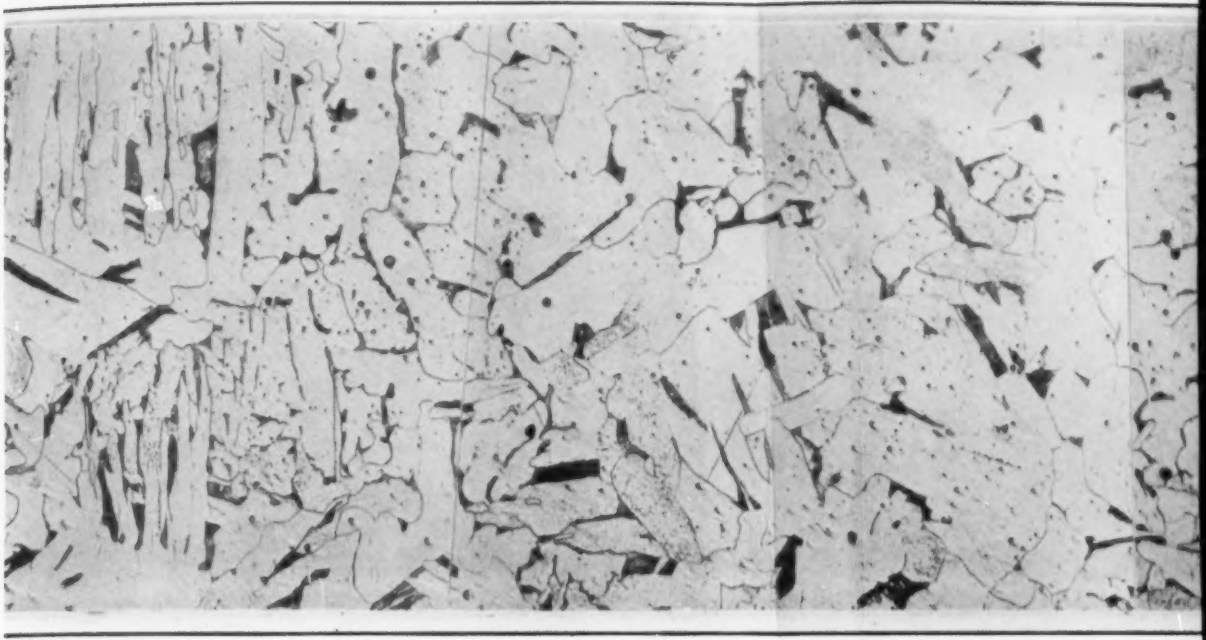
that satisfactory fusion has resulted in cases such as the above and a micrographic examination revealed a very definite carburization for a considerable distance into the body of the chaplet.

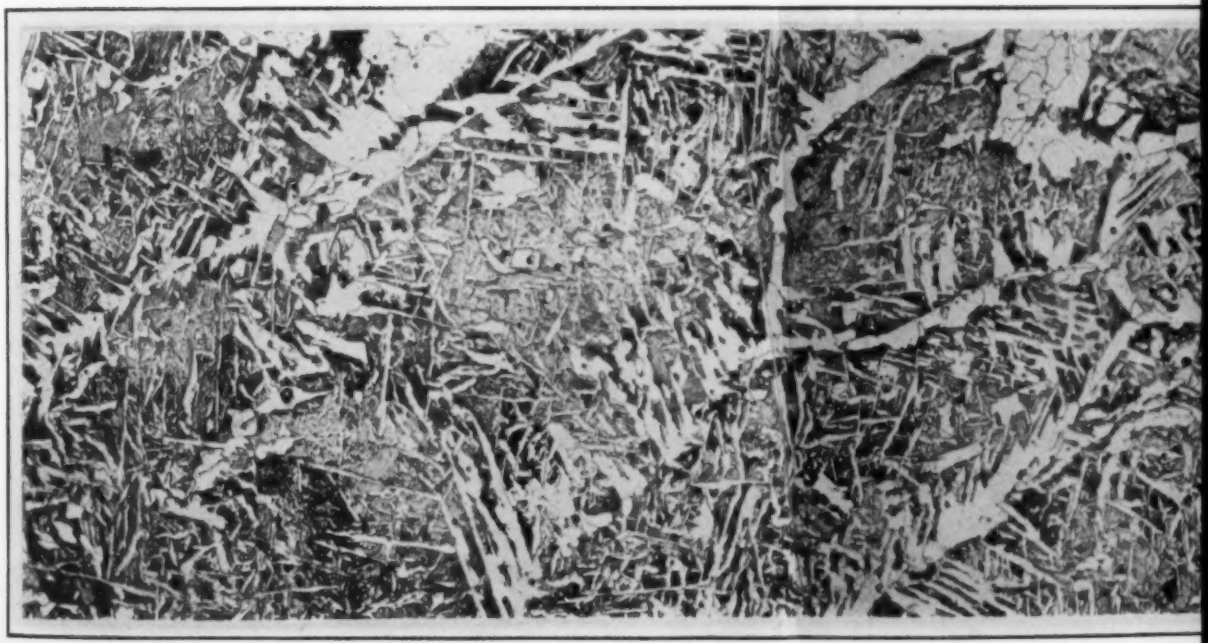
38. Since the above mechanism prevails in the case of cast iron, it is logical to expect that a similar phenomenon exists in the operation of chaplets in steel castings. Steel is poured with enough superheat to melt a low carbon chaplet material but in cases of thin sections, or when the metal is forced to run long distances, a large amount of this heat is dissipated to the sand. As a result, many borderline cases exists where temperatures are possibly too low to effect satisfactory fusion, unless some carburization operates jointly to raise the surface carbon content of the chaplet material. That a certain amount of carbon penetration does take place is shown clearly in Fig. 15. The composite photograph shows the polished and etched surface from the center of a chaplet outward into the body of cast metal.

39. The increasing amount of pearlite from the center of the chaplet to the interface is an indication of the carbon penetration, as is the increasing ferrite noticed from the unaffected cast metal toward the fusion line. The carburization in the present case is of a low order and the difference in melting point between the original compositions of chaplet and casting is slight. In cases of low or medium carbon steel castings, little advantage from the standpoint of a lowered melting point would result. In higher carbon castings, however, it is probable that the influence of carburization would operate to advantage.

40. Specimens selected as representative of good and poor fusion, threaded and superstem types, both in longitudinal and transverse cuts, were carefully polished, etched and examined under high magnifications to see if any conclusions made from the macrographic study should be modified or changed.

41. Considering the design of the coarse thread type of chaplet stem, it can be said that there is a definite trend toward increased porosity at the base of the "V" or, in other designs, at any indentation or sharp internal angle. This is probably the result of dust or moisture having collected at these points before use. Also, such places would be ideal for trapping and localizing any gases that might form at or near the interface. Some cases of good fusion showed no apparent difference between the points





CAST METAL

FIG. 15—ST

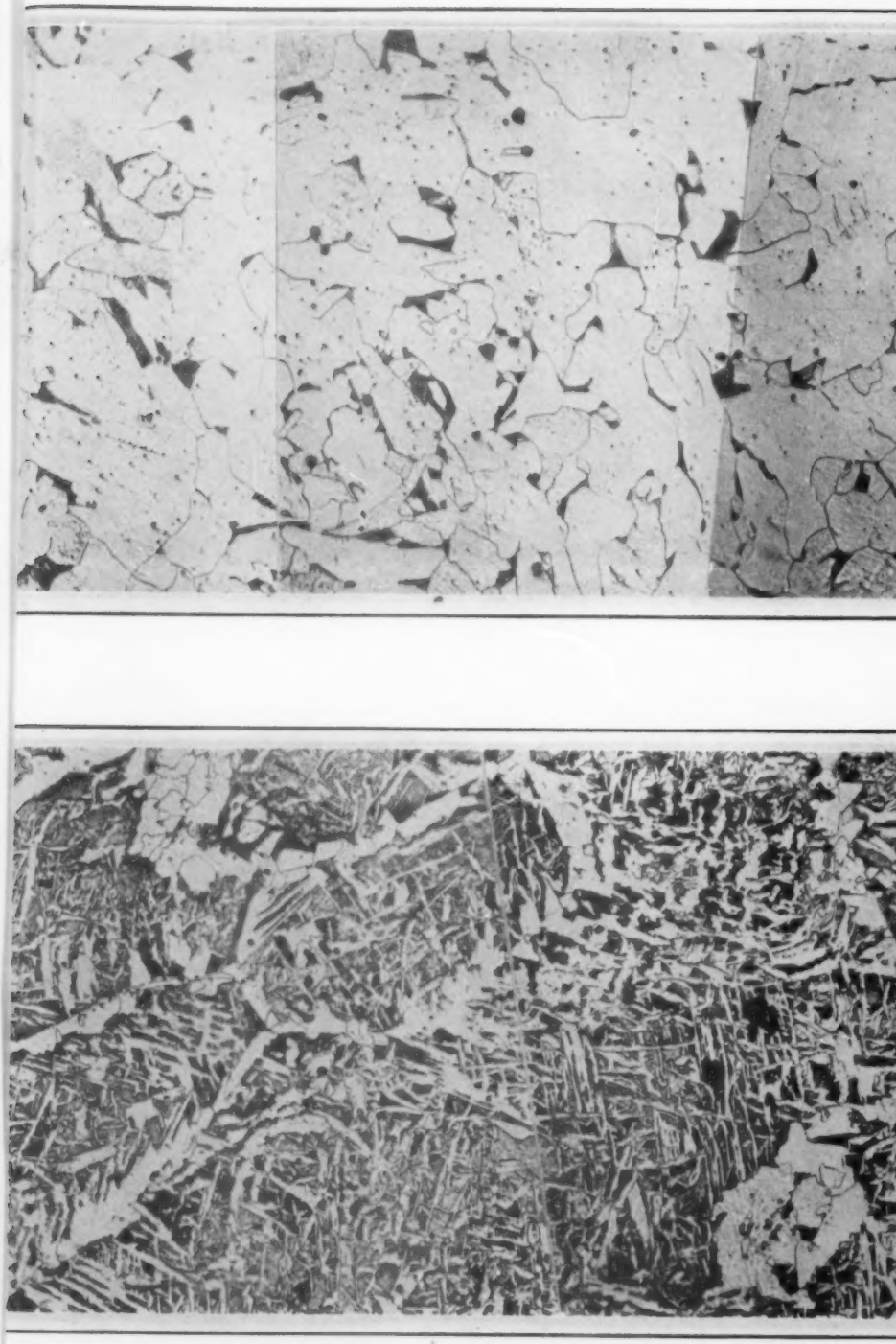


FIG. 13—STRUCTURE FROM CENTER OF CHAPLET INTO THE CAST METAL.

CHAPLET MATERIAL

FUSION ZONE

UNIV. OF MICH.

or bottoms of the "V's," but there is no doubt that a definite keying action does take place in porous specimens. The sharp points of the threads fuse into the cast metal and may appear perfectly sound under qualifying pressure tests. Radiography, if carefully done, may reveal the porosity.

Part 2 — Tests of Specially-Prepared Chaplets

42. The results of the initial investigation, presented in Part 1, were sufficiently encouraging to merit further study and, as mentioned previously, this was done on specimens prepared and handled under carefully controlled laboratory conditions. The basic material, in the form of bare, uncoated chaplets in a variety of shapes and sizes was again donated by the company which supplied the chaplets for the investigations recorded in Part 1. These were of the coarse thread, superstem, and plain stem styles of both ¼ and $\frac{3}{16}$-in. stem diameter.

EXPERIMENTAL PROCEDURE

43. The work was laid out and chaplets prepared in such a way as to answer (or at least indicate a trend) the following questions:

1. What, if any, is the influence of chaplet cleanliness before plating or dipping? Should the base material be sand-blasted before coating, or would a simple electro-cleaning treatment, as normally practiced, be sufficient to remove all deleterious surface scales or films?

2. What would be the difference in efficiency between chaplets prepared under ideal conditions and kept in a desiccator and similar ones allowed to remain on the foundry floor free to collect dust and moisture?

3. What would be the difference in fusion characteristics between the ¼-in. and the $\frac{3}{16}$-in. stem diameter types placed in identical metal sections and poured at the same metal temperature? Would the smaller show any marked advantage?

4. What would be the relative abilities of the various types of coatings to withstand the corrosion normally encountered in storage and use?

5. What would be the influence, if any, of metal analysis and state of deoxidation of the cast metal on the fusion of badly corroded or unclean chaplets? In other words, could a certain amount of poor chaplet reaction be blamed on steel composition?

6. What further evidence can be obtained regarding the influence of stem design?

7. What are the relative merits of different possible coatings weighed in the light of experimental and practical values?

44. To shed light on the above questions, the bare chaplets, as received from the manufacturer, were treated in series as follows:

1. ¼-in. coarse thread style — sandblasted — electrocleaned, coated, and kept in desiccator until used.

2. ¼-in. coarse thread style — electrocleaned, coated and kept in desiccator until used.

3. ¼-in. coarse thread style — sandblasted, electrocleaned, coated and left on the foundry floor free to collect dust and moisture.

4. ¼-in. plain stem style — sandblasted, electrocleaned, coated and kept in desiccator until used.

5. $\frac{3}{16}$-in. plain stem style treated same as 4.

6. ¼-in. superstem style treated same as 4, but placed in a washed green sand mold prior to drying and cast in a high silicon steel.

45. In treatments 1 to 5 inclusive, the following coatings were employed—silver plate, silver + cadmium plate, cadmium plate, copper plate, nickel plate, tin dipped, calorized, and aluminum dipped. In treatment 6, however, the cadmium plated piece was eliminated on the basis of past behavior and an aluminum sprayed chaplet substituted.

46. The pouring procedure was identical to that of Part 1 and the temperature of pouring 2900°F. This time, however, in the interests of constant conditions, the eight specimens shown in the figures were placed in a single mold. Two were equally

spaced in the same relative position in each arm of the test piece that 1 had occupied in the earlier work. Fig. 5 is a photograph of the actual mold used in the experiments, with the chaplets in place. This particular mold was washed and dried whereas all other series were poured in green sand molds. The analysis of the metal for the work shown in Figs. 16 to 25 inclusive, was 0.21 per cent carbon, 0.65 per cent manganese, 0.34 per cent silicon while the series shown in Figs. 26 and 27 were from a high silicon steel containing 0.17 per cent carbon, 0.62 per cent manganese and 1.15 per cent silicon.

DISCUSSION OF RESULTS

47. From a comparison of Figs. 16 and 18, there seems to be a trend in favor of the specimen being sandblasted before coating. Whether this is the influence of metal scale or some film built up during forming the chaplet stock, which is not removed by electrocleaning is not known, but it is certain that there exists a very noticeable difference in the appearance of the two surfaces. In the case of the tin dipped chaplet, the fusion was better in the piece which was not sandblasted. Either extreme condition could prevail, depending on the surface condition and previous treatment.

48. Figs. 16 and 20 afford a comparison to be made between chaplets kept ideally clean and ones similarly prepared but exposed to conditions of foundry storage. It must be noted that these conditions were not nearly as aggravated as those usually encountered in the average foundry since they stood exposed only for a matter of 3 weeks and the foundry was not extremely dusty. However, there seems to be enough difference to indicate that cleanliness and care in storage are matters for attention. Several of the chaplets tarnished very noticeably in the time allowed.

49. In Figs. 22 and 24 can be seen the relative influence of chaplet stem diameter. By paying special attention to the longitudinal sections cut lengthwise of the chaplet stem, there is an overall better fusion of the smaller type. At the same time a glance at Fig. 25 shows that probably a satisfactory amount of unmelted chaplet remains to support a core. However, the difference in fusion is slight, as was shown to be the case in Part 1.

50. It is fortunate that this latitude exists. What might be the case of metal poured markedly colder or hotter can only be postulated. The etched examples tend to show that the smaller

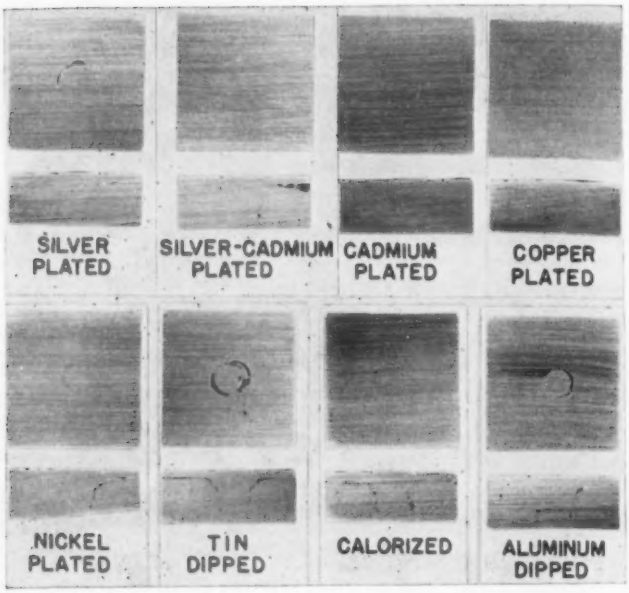


FIG. 16—INFLUENCE OF SURFACE COATING ON ¼ X 1 X 1¼-IN., SQUARE-HEAD, THREADED STEM CHAPLET, SANDBLASTED BEFORE COATING AND KEPT IN DESICCATOR. POLISHED.

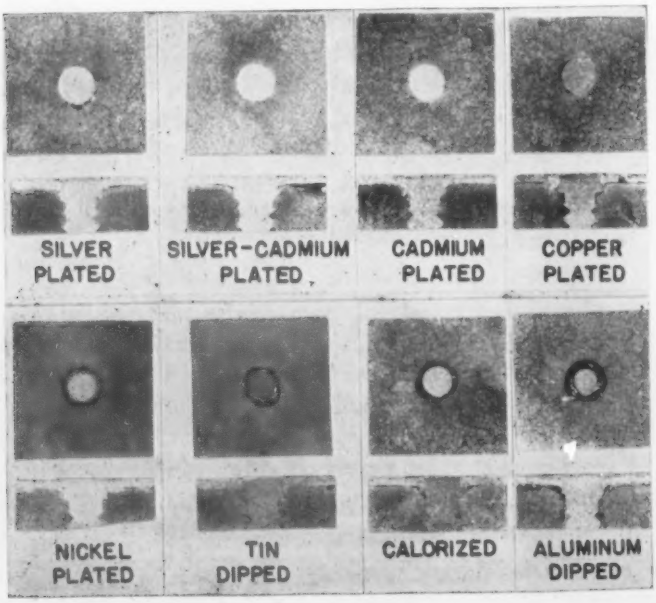


FIG. 17—SAME AS FIG. 16 BUT ETCHED WITH 3 PER CENT NITAL.

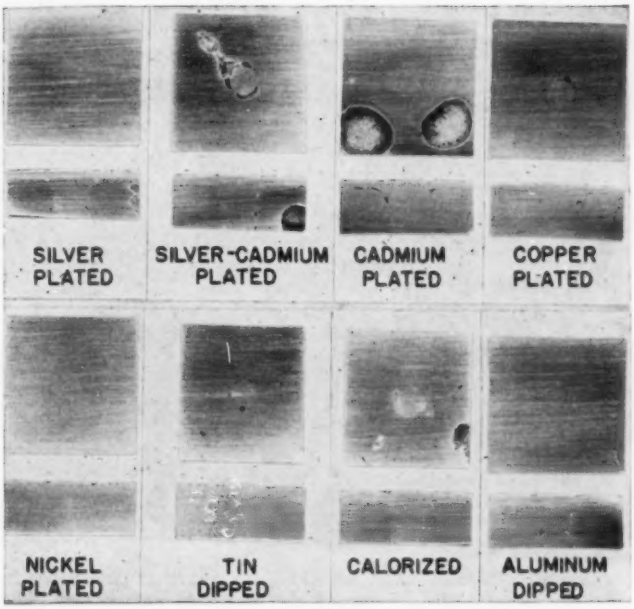


FIG. 18—SAME AS FIG. 16 BUT PREPARED AS RECEIVED (NOT SANDBLASTED) AND KEPT IN DESICCATOR. POLISHED.

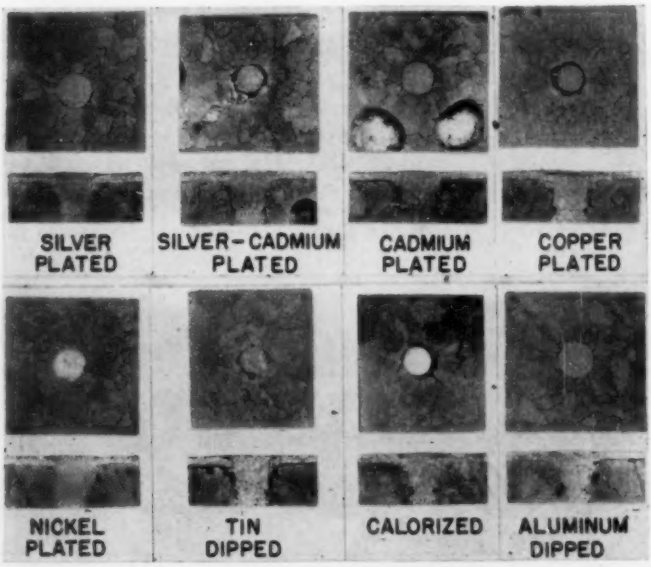


FIG. 19—SAME AS FIG. 18 BUT ETCHED WITH 3 PER CENT NITAL.

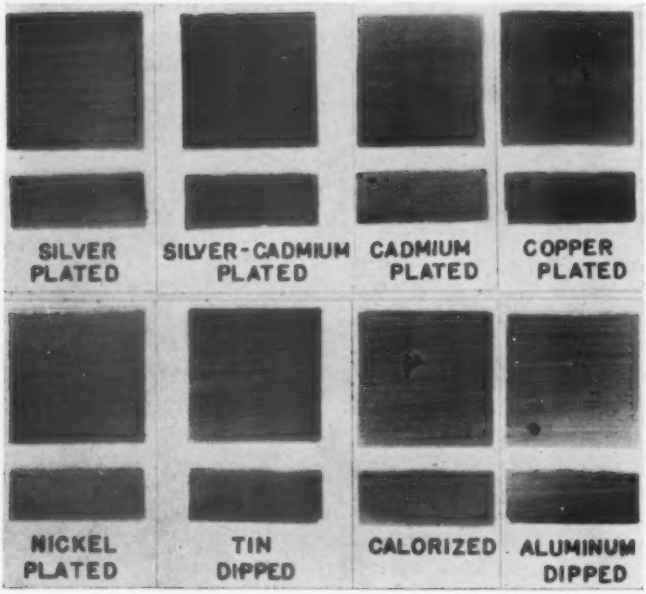


FIG. 20—INFLUENCE OF SURFACE COATING ON ¼ x 1 x 1¼-IN., SQUARE-HEAD, THREADED-STEM CHAPLET WHICH WAS SANDBLASTED BEFORE COATING AND KEPT IN THE FOUNDRY. POLISHED.

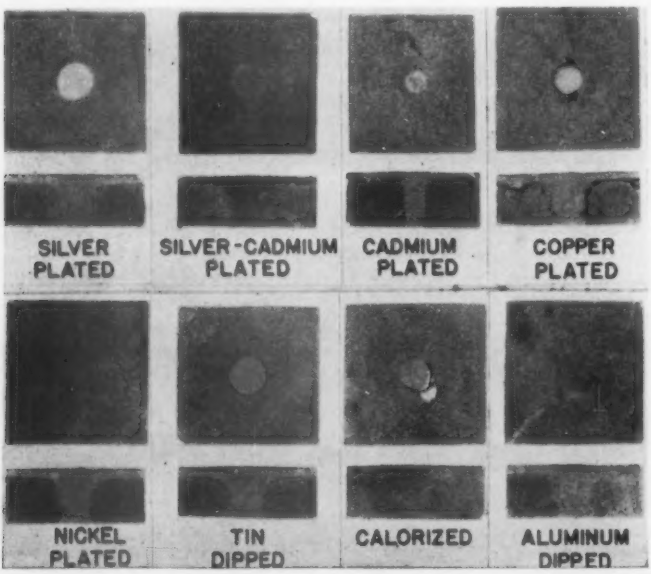


FIG. 21—SAME AS FIG. 20 BUT ETCHED IN 3 PER CENT NITAL.

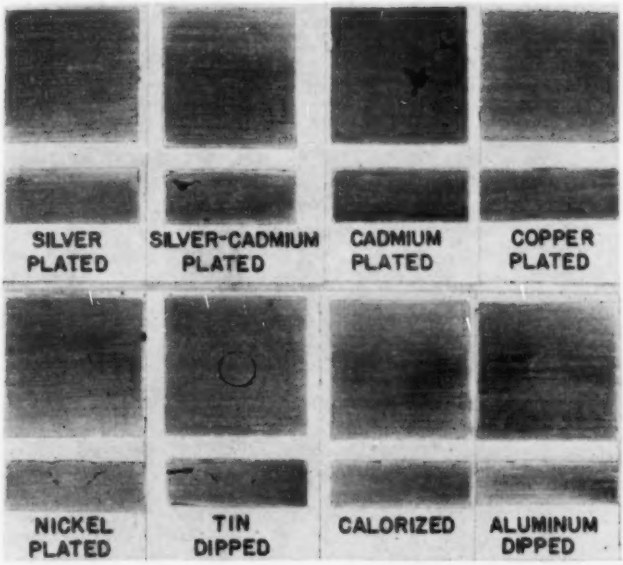


FIG. 22—INFLUENCE OF SURFACE COATING ON ¼ X 1 X 1¼-IN., SQUARE-HEAD, PLAIN-STEM CHAPLET WHICH WAS SANDBLASTED BEFORE COATING AND KEPT IN DESICCATOR. POLISHED.

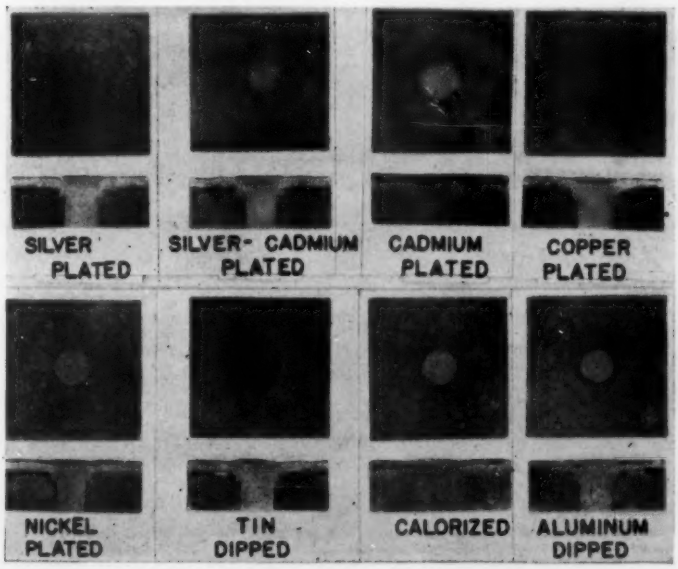


FIG. 23—SAME AS FIG. 22 BUT ETCHED IN 3 PER CENT NITAL.

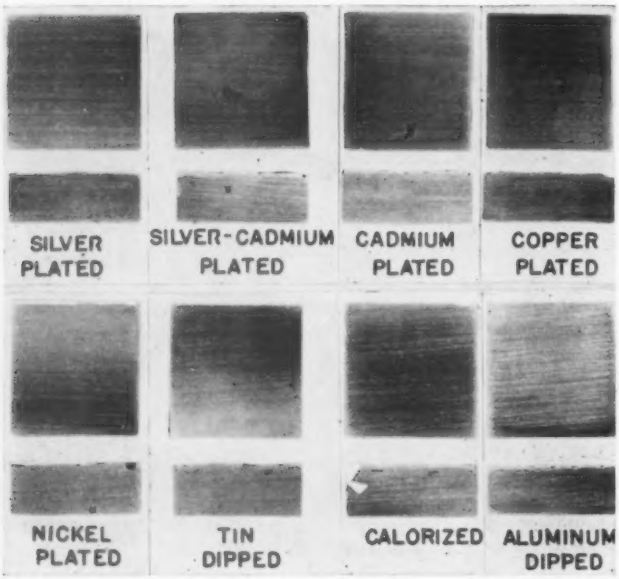


FIG. 24—INFLUENCE OF SURFACE COATING ON $\frac{3}{16}$ X 1 X 1-IN., ROUND-HEAD, PLAIN-STEM CHAPLET WHICH WAS SANDBLASTED BEFORE COATING AND KEPT IN A DESICCATOR. POLISHED.

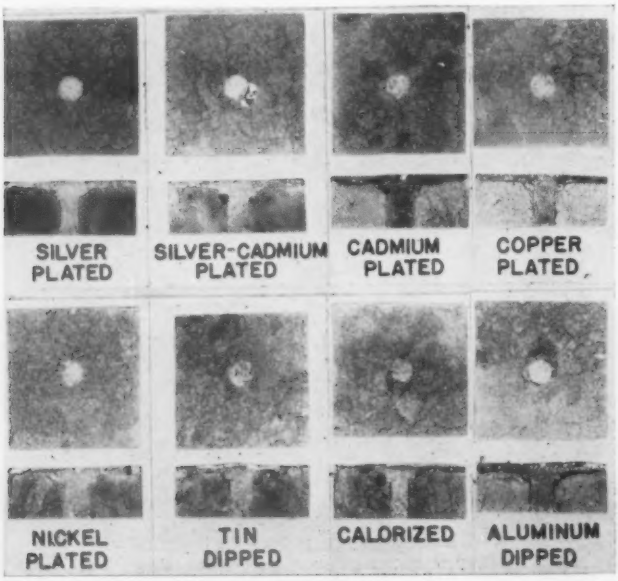


FIG. 25—SAME AS FIG. 24 BUT ETCHED WITH 3 PER CENT NITAL.

chaplet does not disturb the grain structure as much as the larger one but this is no doubt unimportant.

Resistance of Coatings

51. To check the corrosion resisting characteristics of the various coatings, specimens were placed outside the window and subjected to the changing climatic conditions of wind, rain, snow and sun. The series "dried in place" in the washed, green-sand mold gave further evidence. In general, and only from visual estimation, it can be said that the ability of the various coatings to prevent corrosion is about as indicated in the electromotive series of metals. That is, in the order of decreasing resistance, they could be placed aluminum, cadmium, tin, nickel, and copper with silicon impregnation in some position near the top of the list.

52. It must be borne in mind that not only must a coating be corrosion resistant but it must also be continuous. A porous covering can be very deleterious because of rust or moisture accumulating on the unprotected portions. Thus, the care and technique of preparation can play an important part, as well as the density of the coating.

53. It was thought that perhaps a certain amount of chaplet trouble might be the result of the chemistry of the steel poured around it. As a result, a mold was prepared in green sand, painted heavily with a mold wash, chaplets placed in position, and the whole dried slowly in a core oven. This gave the chaplets an aggravated condition for corrosion and there was no doubt concerning the efficiency of the medium. Every porous spot rusted badly and certain of the specimens were completely covered. When used in a casting they would present the worst possible condition. A well-killed steel containing 1.15 per cent silicon (analysis above) was used for the casting. Figs. 26 and 27 show the result.

54. In spite of the apparently bad surface condition of the chaplets, nearly every one fused satisfactorily. Also from Fig. 27, it appears that silicon has diffused from the base metal into the chaplet stem. Such was found to be the case by micrographic analysis. No doubt a better deoxidizing practice would lead to less trouble from chaplets and internal chills. A high silicon heat was used because it was readily available from other work being done at the time and not because it was a practical commercial

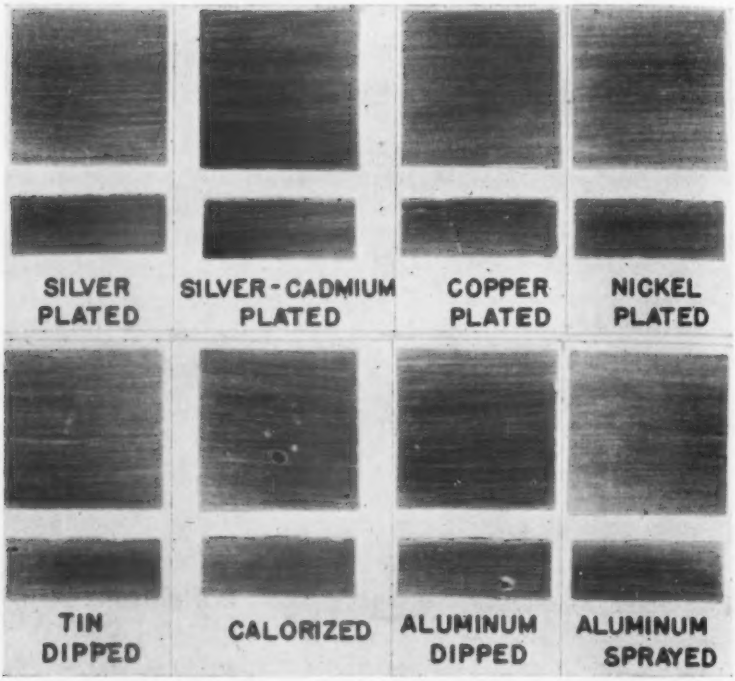


FIG. 26—INFLUENCE OF SURFACE COATING ON ¼ x 1 x 1¼-IN., SQUARE-HEAD, SUPER-STEM CHAPLET WHICH WAS SANDBLASTED BEFORE COATING AND PLACED IN WASHED, GREEN-SAND MOLD PRIOR TO DRYING. POLISHED.

analysis. However, it should be representative of a well-deoxidized steel. Further work using special deoxidizers would be valuable.

55. The effect of the two extremes of stem design, *i.e.*, the coarse thread and the plain stem type are portrayed in Figs. 16 and 22. Under the conditions of these tests, there is very little difference noticeable, but this might be expected since the chaplets were handled under ideal conditions. Therefore, no accumulation of gases or dust would be likely to collect in the depressions. In the plain stem type, there was a tendency for the porous spots to localize at the junction of the stem and head of the chaplet whereas, while the same tendency is usually shown in all cases, there generally appears an increased number of holes near the base of the "V" in the coarse thread type.

Best Coating Problematical

56. It was hoped that the experimental evidence might be

conclusively in favor of one or more of the coatings tried. This would have allowed a classification into some definite order on the basis of practical operation. While the present work does not clear this issue entirely, the results do show very definitely in favor of certain ones over others.

57. For example, it can readily be seen that the silver coated chaplets functioned very well under all the test conditions. It might not however, be enough better than copper, nickel, or aluminum to merit its extra cost. This coating was tried on the basis of the absorbing power of silver for gases. It was thought that perhaps the coating would absorb such gases as seemed to be quite generally present at cast metal-chaplet interfaces. It is seen that a silver-plated chaplet would function properly and would certainly present no problem by solution into the base metal.

58. In the interest of increased corrosion resistance, a specimen was tried with a layer of silver covered by a layer of cadmium, the latter having excellent corrosion resistance. This would

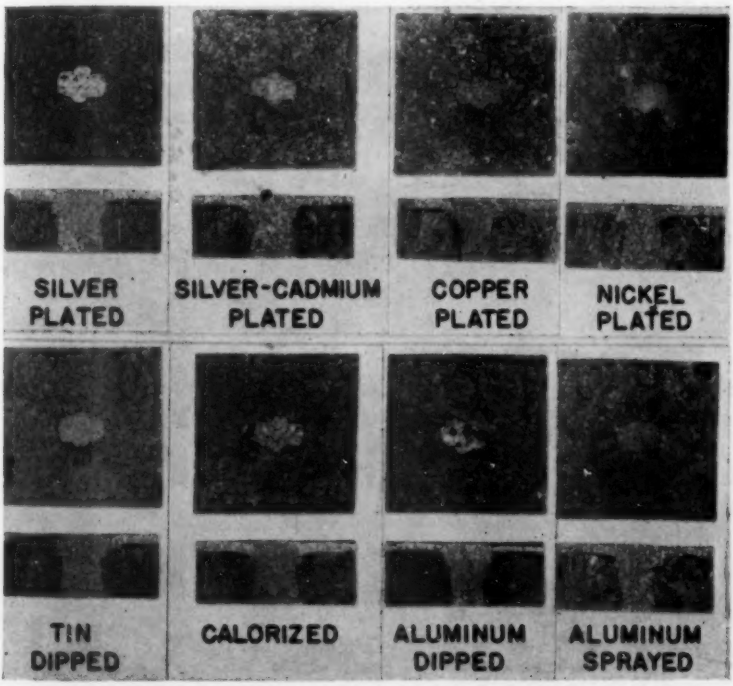


FIG. 27—SAME AS FIG. 26 BUT ETCHED IN 3 PER CENT NITAL.

permit advantage to be taken of the gas absorbing power of silver and at the same time provide a very corrosion resistant coating. This type was not satisfactory because of the influence of the cadmium, which has a high vapor pressure and produces extensive blowholes as a result of its volatilization. In four of the tests, this type was quite good but was not dependable, as two cases were unsatisfactory.

59. From these tests it appears that cadmium is not suitable as a chaplet or internal chill covering.

60. The copper-plated series proved very satisfactory in all tests, except the one employing the specimens which had been kept on the foundry floor to collect dust and moisture. From the standpoint of convenience and economy, this type or the nickel-plated type would be the best choice. Copper-plated surfaces, if not carefully stored, will oxidize to a considerable extent and often because of the porous nature of the coating as plated, some iron oxide will build up. However, with reasonable care in handling, this should provide a satisfactory chaplet.

61. Like copper, nickel seems to give consistent and fully satisfactory results. The choice of chaplet material and cleanliness of surface, as well as stem diameter no doubt accounts for the failure of some nickel-flashed internal chills to fuse properly. Perhaps a certain amount can be laid to the condition of the steel.

62. Contrary to the results of the work on the commercially plated specimens, the tin dipped chaplets were not consistently satisfactory. There is some reticence on the part of many foundrymen to use tin since they believe it to be deleterious to the steel. Lacking any definite data on the subject, it is impossible to say what the exact influence might be but, on the basis of general observations, one is lead to believe that what small amount would be washed from chaplets would not seriously effect the physical properties of the casting. In trials at various foundries, tin was not found as suitable for a coating as copper in regard to fusion characteristics.

63. Aluminum dipped and calorized chaplets were not sufficiently good to merit the extra expense in preparing. Aluminum dipping is not an easy matter and such a process has not been commercially developed. It would seem that the calorized (aluminum impregnated) chaplet would fuse rather readily in the manner of

the silicon impregnated variety. However, such was not the case but the layer of aluminum rich iron was extremely thin. Further work on the aluminum impregnated variety will no doubt prove of value.

64. Aluminum sprayed chaplets were prepared by thoroughly sand blasting and spraying with a metallizing gun. It is imperative, in this instance, that the base material be perfectly free of dirt or grease or the sprayed layer will not adhere. A trial was made with chaplets as received and not sandblasted but the sprayed metal would not stick. Figs. 26 and 27 show the only specimens tried which resulted in nearly perfect fusion. While expensive and troublesome to prepare, this type may yet prove of practical importance. Because of its value as a powerful deoxidizer, this type would no doubt function equally as well in steels which were not so fully killed. This type has excellent corrosion resisting properties as well.

SUMMARY AND CONCLUSION

65. The work presented in this paper has been done in spare time between other investigations and lacks completeness because of this. Nevertheless, it is felt that a certain amount of information has been gathered of value to foundrymen and certainly a type of test piece is presented which should be easily employed for further work in the commercial shop.

66. Attention has been called to many minor points in the choice and handling of chaplets not normally thought important but often it is the little things that count and a saved casting, or an obviated job of chipping and welding, represents big dividends for a little forethought and care.

67. Certain suggestions are made for the purpose of stimulating constructive thinking. The complete lack of experimental information in the technical press on the subject of chaplets indicates that some such attention is definitely needed. There has been a general tendency toward the elimination of chaplets and internal chills to as great a degree as possible. Is it not probable that basic research and general shop testing would eventually lead to a chaplet material and design which would function satisfactorily in nearly all cases and remove in large part the stigma against them?

68. Remembering again that chaplets are only a specialized

form of internal chill, it follows that the development of a suitable coating for the former would provide an important tool for the treatment of many solidification problems. Internal chills, giving good fusion at a corrosion free surface, would enable their use to great advantage in those cases where poor design or necessity demands it.

69. The conclusions which can be formed from this work will vary considerably with the individual viewpoint regarding the importance of the issue. However, assuming that the smallest detail merits consideration in a field far too little investigated, the following observations are made:

1. Many generally ignored factors operate jointly or individually to cause unsatisfactory operation of chaplets.

2. Threaded stem chaplets do not present any added assurance of satisfactory fusion and may, in some cases, be deleterious by presenting spaces at the bottom of threads for the accumulation and localization of dust, moisture and subsequent gases. Threaded stem types do promote a keying action by fusion at the points of the threads. This may be an aid in passing a qualifying test, but without complete fusion at the base as well as the point, the casting may fail in service. Perhaps the square stem type twisted torsionally might give this desired keying action at the same time being stronger and obviating the sharp V's.

3. More consideration should be given to the storage and handling of chaplets.

4. Too heavily alloyed chaplet material fuses readily but may sometimes melt too easily for suitable core support. A properly-chosen thinly-impregnated case would promote the necessary fusion and yet the base material would probably not melt too readily.

5. A good grade of low carbon steel is well chosen for chaplet material to be used in steel castings.

6. Fusion is aided greatly by carbon diffusion from the higher carbon base metal into the chaplet material at the interface. This diffusion lowers the melting point of the chaplet and promotes satisfactory fusion at a clean surface.

7. Silicon impregnated chaplets fuse readily. A certain amount of grain coarsening results in the casting for a small distance around the imbedded chaplet. Perhaps this type would operate more satisfactorily with a thinner case than used in the present tests, as mentioned in 4.

8. While the tin dipped variety proved good in Part 1, further tests did not bear this out consistently.

9. Copper and nickel-plated chaplets proved satisfactory when prepared properly and kept clean.

10. Silver-plated chaplets were the most consistently satisfactory of any series tried. Silver-cadmium and cadmium coatings proved unsatisfactory because of the high vapor pressure of cadmium.

11. Aluminum sprayed chaplets are thought to be very good but too little actual data was collected on this type to allow an accurate statement to be made. More work on the calorized and dipped types might prove of great value.

12. So-called streamlining of chaplets would obviate the tendency of gases to form and localize in indented areas.

13. The duplication of test results indicate the suitability of the testing procedure used.

ACKNOWLEDGEMENTS

70. It is with pleasure that grateful acknowledgement can be made herein to the following men who aided in this work: T. Cunningham, J. Darby, A. L. Donaldson, and I. R. Kramer. Also to the Navy Department for sponsoring the work and for permission to publish the results.

DISCUSSION

Presiding: H. D. PHILLIPS, Lebanon Steel Foundry, Lebanon, Pa.

T. D. WEST[1] *(Written Discussion):* To the foundryman, chaplets are just a support for cores and are handled as such with no regard or thought. However, anyone making high pressure or gamma-rayed vessels can easily see the trouble that thoughtlessly placed chaplets can cause.

The writer would like to add to the authors' list of factors affecting

[1] Foundry Superintendent, Symington Gould Corp., Rochester, N. Y.

chaplet reactions. The reactions would be different in direct relation to the amount of metal passing the chaplet, the rate of flow past the chaplet, the weight that it is to support and the chaplet should be visible after etching or it will not have served its purpose.

In the past, we have run a series of tests on chaplets and the following is a list of our results:

1. The pickled or shot blasted chaplet was the best. This mold was poured very quickly after setting of chaplets.

2. The copper coated one was next best. The copper can plainly be seen as diffusing into the steel upon etching. This undoubtedly gives one a different steel at this locality.

3. Tin coating was not entirely satisfactory.

4. All chaplets showed a defect at the joint of the upper plate and stem. This was thought to be due to either the chilling effect of the two parts or gas trapped there as the metal rose up the stem.

5. The flat sides of the double angle chaplet should run longitudinal, or parallel to the direction of tension, and not at right angles.

6. Chaplets with more metal passing them proved themselves more fused than the ones farther away from the gate.

7. Butt welded stem chaplets showed up a little better than upset chaplets.

8. Chaplets in the horizontal position were fused better than chaplets in the vertical position.

9. The square twisted stem chaplets were the most desirable.

10. It is recommended that chaplet pads be put on all work and chipped off. This eliminates that defect which showed up at the injunction of the top plate and the stem. Or the chaplet can be removed by drilling and welded up.

11. Chaplets poured with hotter steel fused better than those poured with colder steel.

These tests were made on almost every type of chaplet, round and square on stem, angle, reeder, threaded stem, perforated, angle brace, bridge and herringbone stem and box chaplets.

The coatings were as follows: nickel, cadmium, chromium, copper, shellac, varnish, wax, core oil, linseed oil, pickled, shot blasted, electrotin, hot tin and silver coatings.

Tests were run for the determination of size and strength of chaplets needed for different work. These tests have not been completed and are not entirely satisfactory as there are variables which have to be studied first.

This fine paper certainly gives the foundryman much to think about on the subject of chaplets and the writer wishes to compliment the authors and to thank the Navy Department for the release of another paper.

J. A. DUMA[2]: There is one point that is not clear to me. The authors have advanced the theory that the melting point of the chaplet is lowered by carbon penetration. I wonder if the authors have determined whether the carbon penetration was effected after solidification or prior to solidification.

A couple of years ago Briggs and Gezelius, in their skin solidification studies*, showed that metal coming in contact with a cold wall, like a chaplet or the face of a sand mold, forms a skin at a high rate of speed, a matter of a second or so. I just wonder whether that is time enough for the carbon to effect penetration into the chaplet, or whether that carbon has penetrated after solidificaion has taken place. In carburizing practice, we know that it is the long time at high heats that effects carbon penetration.

MR. TAYLOR: We much appreciate this contribution in regard to whether the carbon diffusion or penetration takes place before or after the melting. I would say that, in the work by Briggs and Gezelius, they were concerned with larger sections than we. The metal, coming in contact with a large chill, would certainly form a skin quickly, and in that case we probably would not accomplish much by lowering the melting point by diffusion before the metal had solidified.

However, in the smaller chaplet stem diameters, there is considerable metal that flows past the chaplet stem because we found porosity, indicating to us that the gases formed at the cast metal-chaplet interface had been washed along in the stream. In many cases, these defects show up, and I believe it is due to that. Considering that the case, I would say the carbon diffusion has taken place before the skin was formed.

In larger chill, that would be something else again. In the case of cast iron, if this were the case, if the diffusion had taken place after the metal had solidified, we could not fuse low carbon steel chaplets in cast iron, but it has been done, and so that more or less speaks for the penetration having taken place before solidification occurs.

S. W. BRINSON[3]: I am interested in a technique for holding the floating force of the core after the casting is poured. We all know after a casting is poured, there is four times the weight of the core tending to push up the core. That is where the trouble takes place. Any kind of chaplet can be used to support the core until the metal gets around it, then the chaplet can be thrown away, but as long as the steel

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[2] Assistant Metallurgist, Norfolk Navy Yard, Portsmouth, Va.
[3] Master Molder, Norfolk Navy Yard, Portsmouth, Va.
* *"Studies on Solidification and Contraction in Steel Castings—III The Rate of Skin Formation,"* Briggs and Gezelius, TRANSACTIONS, American Foundrymen's Association, vol. 43, p. 274, (1935).

is molten there is a force tending to push that core to the top of the mold. With too great a fusion, the question is whether there will be strength enough left in the chaplet to hold the core down. If you have ever tried putting an anchor or rod chaplet down a riser on a steel casting you will know what I mean. The heat, as it arises, will affect the chaplet so much it allows the core to come up, and if the chaplet is fused too much you will find there is not enough strength left in the chaplet to hold the core down, and you are liable to get a 'cut through at the top. That is one reason that in cast iron you cannot use a cast iron chaplet because it will fuse too quickly. I think there is a borderline. If you can fuse it, well and good, but you have got to allow for enough in the chaplet, so that after fusion takes place to a certain depth there is still enough steel in the center to hold the core down. Personally, I have tried to eliminate the use of chaplets, as far as possible, using any other way in order to do so. We all know the chaplet is a potential source of defect in the casting.

MR. TAYLOR: I might say that before we have any reason to keep a core from floating we have to support it—we have to keep it from sinking, so we have to have something below as well as something above. Work along these lines will be taken up later.

In this matter of fusion, I doubt if it is absolutely important that we have pronounced fusion, but we do have to have a clean chaplet. We have to have a surface which will not produce gases, whether it fuses or whether the metal just lies quietly up against it. There probably is some sort of reaction taking place right on the surface, which does not penetrate very far, and when you look at specimens under the microscope you do find that there is a slight fusion. That is what I meant by choosing a chaplet with a little precaution and care. *Know your chaplet!* If you use a certain low melting variety, surely it is going to fuse if very much metal flows past it; whereas if you are pouring on the cold side, if your chaplet is a long way from your gate, this type probably is going to be very beneficial. In other cases, you have to choose your chaplet, maybe with a still lower carbon content, so it does not fuse so readily.

E. C. TROY[4]: The use of a silicon impregnated chaplet is very interesting. Since, however, the photographs indicate a migration of the silicon to a considerable depth in the parent metal, is it not possible that there will result a zone in which the steel will contain enough silicon to have a typical brittle and non-heat-treatable condition of the 3 per cent or more silicon steels?

Mr. Duma raises the question of the possibility of carbon migration from the parent metal to the chaplet. I have seen evidence of the reverse migration from internal chill to parent metal. In using (by error) some 0.70 per cent carbon rod for internal chills, there developed a reaction sufficient to badly blow the parent metal. On checking back later, it was found that if a chill of 0.70 per cent carbon was used

[4] Metallurgist, Dodge Steel Co., Philadelphia, Pa.

and any fusion at all took place, regardless of how clean the surface of the chill might be, a reaction occurred.

We assumed this reaction to be the result of a new carbon-oxygen equilibrium in the parent metal.

The suggestion in this paper that some of the non-metallics noted near the chaplet are due to dust on the chaplets forming fused products, seems less likely than that these non-metallics are nothing more than oxidization products of silicon and manganese. The same products can be noted near the edge of the casting at the mold face or generally anywhere large or even small amounts of oxygen are present to upset the iron oxide equilibrium of the liquid metal.

The use of a 1.15 per cent silicon melt gave interesting results. Its success is probably due to the ability of the silicon to hold gas in solution and fix oxygen in the form of silicates. However, such an analysis is at the present time not commercially feasible since most specifications fix the silicon at a lower limit.

We foundrymen are again indebted to the Naval Research Group for a splendid program well executed. I wish to thank and praise them for the information available in this paper.

MR. TAYLOR: At present, this high silicon would be considered abnormal, but in the future it may not be. Silicon is really not as badly behaved as some people think it is. There are steels on the market of 1.25 per cent silicon which are very good steels. I think we are going to become more familiar with them shortly. I will not enlarge upon this, but I have reason to believe that silicon can be run quite high in castings. We have made several tests in the laboratory indicating that there is not a great deal of trouble up to 0.8 per cent; if you vary other elements with the silicon, amounts as high as 1.25 per cent produce excellent steels from the standpoint of fluidity and increased yield ratio.

Regarding the influence of the high silicon in these small localized areas as being deleterious, I do not believe it is going to be so bad because it is very seldom that we ever get a high shock in a small localized area, and I believe that the more ductile, smaller grain size base metal surrounding the spot, would probably absorb the impact and prevent localized cracking.

As far as dust not being influential, I think it is, because quite often you see slag at the chaplet-cast metal interface. That may have been trapped there; I do not know. I presume a goodly portion of it was the result of dust. This dust may have collected at the depression or on the smooth surface and some rust formed, and around that dust particle moisture collected, and so on and on, until we get a lot of dust, and if these are turned loose out in the metal they can form pinholes. The penetration or diffusion of gases into these pinholes can cause the hole to be quite large. Mr. Sims has done quite a bit of work on gases in steel and he has shown that such can be the case.

Non-Ferrous Applications of Top-Pouring Methods

By Arthur K. Higgins,* Milwaukee, Wis.

Abstract

At the 1938 convention of the American Foundrymen's Association, Brisbois and Cartwright presented a paper on the application of riser or top-pouring as applied to both non-ferrous and gray iron castings. In 1939, Scott presented a second paper on this subject dealing with further experiments on the theories and practices advanced by the first authors, particularly with their application to gray iron. The present paper is the third of the series and is devoted to the application of top-pouring to non-ferrous castings such as bushings, pump impellers and pump casings. The author finds the method more applicable to the non-ferrous alloys typified by such alloys as 80-10-10, 85-5-5-5, 75-10-15 than to such alloys as aluminum or manganese bronze. He points out that by using top-pouring on the former alloys mentioned that he is enabled to pour at lower temperatures, which enhances the physical properties of the alloys, produces sounder castings, and reduces the amount of metal needed for gates and risers.

1. The foundry with which this paper deals, has been confronted from time to time, as are most foundries, with castings whose design is such as to cause considerable difficulty in choice of gating and risering methods. A number of such castings had been encountered at about the time of the presentation of the paper by Brisbois and Cartwright.[1]

2. Contrary to the views expressed in that paper, considerable trouble had been experienced with such alloys as 80-10-10, 75-10-15 and 88-8-4, rather than with the alloys having higher shrinkage. Difficulty had been experienced with a group of large

* Allis-Chalmers Mfg. Co.

[1] Brisbois, C., and Cartwright, A. E., *"Risers or Gates for Some Special Non-Ferrous and Alloyed and High Test Iron Castings,"* Transactions, American Foundrymen's Association, vol. 46, pp. 219-256 (1938).

Note: This paper was presented at a Non-Ferrous Session of the 44th A.F.A. Convention, May 7, 1940, Chicago, Ill.

80-10-10 bushings of from 500 to 1000 lb. weight, which were of sufficient diameter and length to prevent adequate feeding of central portions when cast with the axis in either a vertical or horizontal position. While feeding might have been secured by an increase in pouring temperature, it was not felt that the added temperature was desirable from the standpoint of obtaining the desired physical properties.

3. The timely presentation of the paper mentioned above, offered the solution to many of the problems confronting us.

PURPOSE

4. It is not the purpose of this paper to retrace the work done by Brisbois and Cartwright, but rather to make more feasible the application of their methods to general foundry work, by the elimination of some of the practices that may have caused difficulty in the smaller foundries that are confronted with the same problems.

5. Points in the method that caused the greatest hesitation in applying the practice to our own foundry were:

(a) The heavy risers shown,

(b) The use of synthetic sands to resist erosion by falling metal and,

(c) The added cost of molding runner boxes and strainer gates for medium and small castings.

The principal work done on top gating has been in attempting to circumvent these difficulties.

PRELIMINARY TESTS

6. The first steps in the application of gating through risers were the production of a number of test castings which were intended to show; (a) the effect of impact of falling metal on dross formation; (b) the effect of rate of pour on shrinkage and feeding; and (c) the ability of natural sand molds to withstand the cutting action of falling metal.

7. It was felt that the first two questions could be answered by casting simple sticks molded on end, and using a high-shrinkage, drossy alloy. The results of some of our tests are shown in Fig. 1, which represents the upper halves of three, 3 x 18-in. manganese bronze sticks.

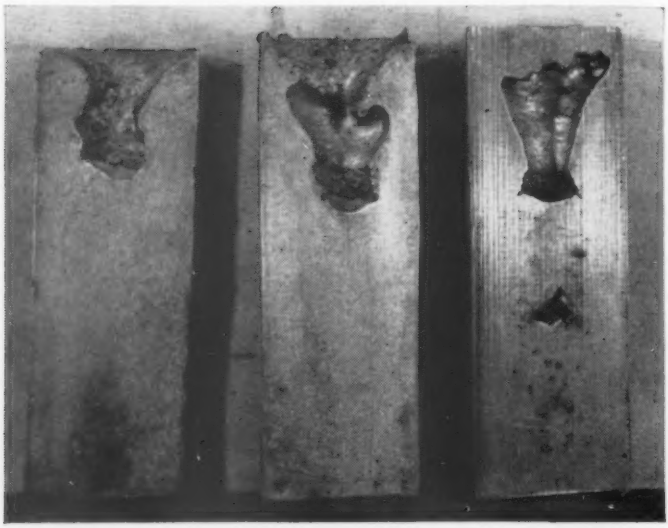

FIG. 1—UPPER HALVES OF EXPERIMENTAL MANGANESE BRONZE STICK CASTINGS.

8. The casting at the left in Fig. 1 was poured from the top through a single ¼-in. hole in a strainer core, and was filled in 115 sec. with metal at 1950°F. The mold was made of No. 1 Albany sand, and no precautions were taken to prevent cutting, except to see that the mold was vertical.

9. Careful inspection of this casting showed that while shrinkage was small (2½-in.), dross was entrapped by the rising metal near the midpoint of the casting.

10. The center casting in Fig. 1, was molded in the same way, except that the metal was introduced through ten ¼-in. holes in a strainer core at the top of the casting. The casting was filled in 15 sec. with metal at 1950°F.

11. Inspection shows that this casting, while having slightly deeper shrinkage (3¼-in.) is sound throughout the rest of its length.

12. The conclusion drawn from the action of these two castings, and others like them, was that excessively slow pouring rates, while reducing shrinkage somewhat, were not desirable from the mechanical standpoint of dross and dirt entrapment. Such slow pouring reaches a critical point where the rate of pour is slow enough to allow the zone of turbulence from the falling metal

stream, to enter the zone where interlacing metal crystals are forming and that the resultant entrapment of dross and accidental dirt may cause difficulty with any type of metal.

13. A stick casting of the same type was poured by the conventional method of bottom gating, with a choke in the gate a little removed from the point of entrance of the metal into the mold, to allow the metal to enter quietly and without turbulence. This casting is shown at the right in Fig. 1. As may be observed, the casting shows open shrinkage to a depth of 5½-in. and shrinkage porosity through nearly 9-in. of the 18-in. length.

14. As a result of these and similar tests, it was concluded that:

(a) We could use top or riser pouring for our castings at normal pouring rates and that excessively slow rates were not only unnecessary but dangerous.

(b) We could pour our common brasses and bronzes in this manner without great fear of drossing in simple castings, and

(c) That we could materially decrease the size of risers without having shrinkage extended into the castings.

Also indicated by these tests, not shown in the illustration, was that a rather rough surface may result at times in castings of aluminum bronze or manganese bronze. This makes gating through risers impractical for these alloys, even in the simple castings, where it could otherwise be used, unless outer appearance is not important, as in fully machined castings.

APPLICATION

15. The first foundry application of this method was in the casting of large bushings of the type shown in Fig. 2, ranging in weight from 600 to 1000 lb. Normal practice had been to cast these bushings with the axis horizontal, to place a riser at each end of the bushings, and to supply the metal through gates at the bottom. Frequent losses were encountered in the larger sizes, due to shrinkage along the parting line. Bushings cast with the axis vertical and gated through the bottom, were also subject to this type of shrinkage.

Bushings

16. The first top gated bushings were poured in dry sand with the use of rather elaborate runner boxes designed to supply the metal through a number of risers by means of skim gates. It has been since found that green sand is suitable for the smaller sizes, if certain precautions are taken to absorb the first shock of falling metal.

17. A practice is made of placing a small flat stock core against the bottom of the pattern at the point where the falling metal will strike when the mold is completed. The first metal entering the mold forms a pool that cushions the fall of the remaining metal.

18. The limitations in using green sand for bushings seems to be the same as for castings with conventional gating, provided the core sand pieces are placed as described in the previous paragraph. Where such cores have not been used, we have had occasional losses due to sand in the castings, when bushing heights have been more than 20 or 30-in.

19. The sand used for all green sand molds is the same as that used in the preliminary castings, No. 1 Albany, and has a green strength of about 4 lb. per sq. in. and a permeability of 18.

FIG. 2—METHOD USED TO CAST 80-10-10 BUSHINGS IN DRY SAND.

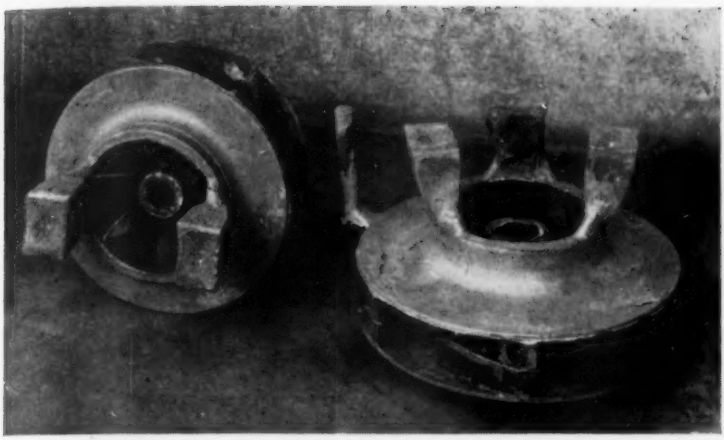

FIG. 3—CENTRIFUGAL PUMP IMPELLERS WHICH WERE POURED THROUGH RISERS. THE CASTINGS ARE POURED WITH 90-10.

20. Bushings of the larger sizes are normally made in dry sand and without special precaution. The one illustrated in Fig. 2 is an example. In all cases, metal is supplied to the mold cavity through a runner box.

21. Runner boxes for the larger size castings are built in green sand but are commonly supplied with flat stock core bottoms, through which holes are filed to admit metal to the mold cavity. Care is taken that the holes are directly over the casting so that the metal may have a free fall.

22. The second application of top-pouring, and perhaps the most important, is in casting centrifugal pump impellers, two of which are shown in Fig. 3.

Centrifugal Pump Impellers

23. The bulk of these castings are made in green sand with a cereal bonded core. The metal, usually 90-10 or 75-10-15 is dropped onto the upper surface of the core, which, as may be seen, is intricate but usually rather massive. Impellers of the size shown, which weigh about 300 lb., may be cast in dry sand molds.

24. Losses experienced in these impellers, when cast by conventional methods, were due to several causes: (a) The heavy outer hub section was subject to shrinkage, particularly on the bottom side where it was cut off from feeding by the relatively thin vanes, (b) in the smaller impellers, having heavy central hubs, the

vanes often cracked away from the hub unless excessive pouring temperatures were used, (c) excessive pouring temperatures caused a great deal of trouble by burning in cores and, since some of these castings have only ⅜-in. opening for core removal, costs were high.

25. Normal gating practice for this type of casting was the use of a loop-gate attached to the upper edge. If the impeller was large, two such gates were used and the castings were poured "double up."

26. Present practice is as shown. Two or three risers are connected to the outer hub and a small flow-off is cut from the outer edge of the mold to indicate the height of metal in the mold. If the center hub is excessively heavy and likely to cause cracking of the vanes, the upper portion of the center core is cut off and a full riser is placed at that point.

27. Metal is supplied through a prefabricated, core-sand, runner box, and is poured into all risers. Pouring is gauged by the flow-off and is stopped while there is room in the risers for all of the metal in the runner box.

28. We have found but one cause of trouble that can be ascribed to top-pouring in this type of casting. That is, in the possibility of cutting out sand from the upper face of the mold where metal bounces off the core. In every case, examination of the surface of the casting has shown that the mold was not rammed as hard as it should have been, since this cutting action is associated with small swells in the casting.

29. Pouring temperatures for these castings are decreased by the top-pouring practice, as was the case with bushings. Where the normal pouring temperature for 90-10 bronze, with the old style of gate, was between 2150 to 2250°F. for castings of 20 to 50 lb. weight, our present practice is to pour these castings at from 2050 to 2100°F. The lower pouring temperatures permit running of the extremely thin edges of vanes and still maintain a good metal structure and good physical properties.

Pump Casings

30. The casing shown in Fig. 4 is a typical example of a type of casting that has caused a great deal of trouble because of pinhole leakage, almost invariably due to shrinkage. The many section

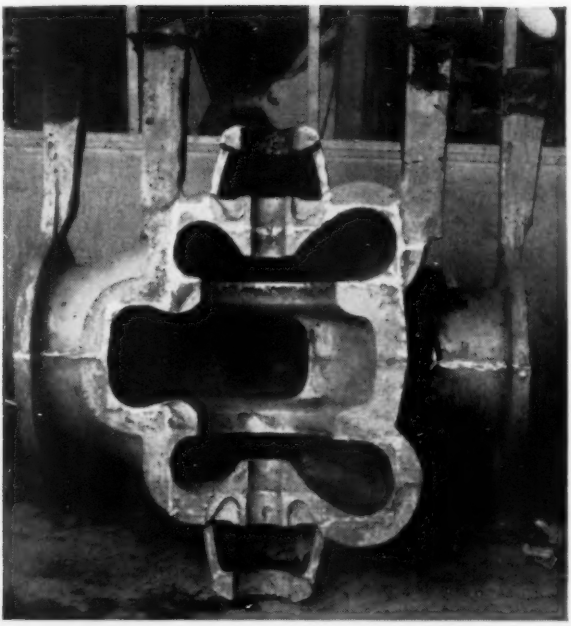

FIG. 4—3500 LB. 85-5-5-5 PUMP CASING POURED THROUGH RISERS.

changes and the frequently severe test specifications have combined to cause trouble in many cases.

31. These pump casings have varied so in size and complexity that no entirely standard method of conventional gating could apply. In general, however, the method has been to connect several gates to the lower parts of the casting and to fill these through a common runner box.

32. Adoption of the new system has resulted in definitely smaller losses, and in improvement in accepted castings. These castings are all made in dry sand and are normally molded in the position shown. A riser is attached to each flange at a point where it will best feed the heavy sections. Metal is poured into the risers through a green sand runner box and is allowed to fall freely into the mold without great regard for the mold and core surfaces. There have been no indications of cutting-in, in cases even more severe than the example shown, where about 3000 lb. of metal fell into the mold before the rising metal covered the point of impact on the cores or mold.

DIFFICULTIES ENCOUNTERED

33. While no insurmountable difficulties have been found in the application of top-pouring to foundry practice, certain factors inherent in the method must be noted. The first of these is the increased molding cost resulting from the use of runner boxes for all top-poured castings. While complicated runner-boxes may be justified by large castings, the bulk of foundry work will not carry the extra molding cost, although the castings made may be of much better quality. Our solution to this problem has been the construction of core-sand runner-boxes that provide for almost all of our small castings.

34. The core box, and the cores made from it, are shown in Fig. 5. The core box is provided with ample draft so that the center piece may be fastened to the bottom of the box. The cores are made from any of a number of heavily bonded core sands and are made in either of two sizes carried as stock. As may be seen from the illustration, the core box does not provide for the holes that admit the metal to the mold. These holes are cut by the coremaker while the core is still green and allow enough flexibility in the runner cores to make the method successful.

35. These runner cores are carried in stock until they are too worn for further use. Normally they are re-blackened after each use and serve from 6 to 8 castings before they must be discarded. It is the duty of the coremaker, assigned to any job, to see that there is a runner core in stock that will fit the casting on which he is working, or to make such a core.

FIG. 5—CORE BOX AND CORE-SAND RUNNER-BOXES FOR SMALL CASTINGS.

36. It has not been the practice to use strainer cores to control the flow of metal to the mold as it is felt that the use of such cores involves needless expense in setting and, in addition, creates a possible source of dirt or sand, particularly in the case of large castings. It would seem as valid to worry about the cutting of a strainer core by metal flow as to worry about cutting of cores within the casting, particularly when the strainer core must carry the weight of the column of metal above it, while the internal core carries only the bouyancy of its own displacement. It has been the practice to so proportion the core-sand runner-boxes for the small casting, or the green-sand boxes for larger castings, that the amount of metal supplied by the pouring basin is more than adequate to keep the down-gates filled. We have made these down-gates for large castings by cutting holes of the desired size through a flat stock core, which serves also as a bottom for the extremities of the box.

37. The second factor of importance in the use of top-pouring is the necessary revision of pouring temperatures. Experience would seem to indicate that the alloys to be cast should be divided into two classes: (1) The normal alloys, which are those having the following properties: (a) little tendency to form dross, (b) great increase in crystal size with moderate increase in freezing time, and, (c) intra-crystalline rather than bulk shrinkage; (2) the so-called high shrinkage alloys that are characterized by the properties of (a) the formation of large amounts of dross; (b) the presence of iron or of some other alloy that serves to control the crystal size, and (c) bulk or inter-crystalline shrinkage.

Pouring Group 1 Alloys

38. The first group of alloys may be represented by 90-10, 80-10-10 or 88-8-4. It has been our experience that such alloys must be poured at lower temperatures when top-gating is used than when they are cast by conventional methods. This decrease in temperature is made necessary by the fact that the metal entering the mold is at a higher temperature than when it is compelled to supply heat to dry and warm the lengthy passage through the mold that constitutes the conventional gate. As an example, a temperature drop was found between the gate and the riser of a 3500 lb. ring, cast by the conventional system, of 180°F., while the drop in the case of a top-poured casting is less than the error of the pyrometer used.

39. The most apparent effect of these higher metal temperatures is the greatly increased crystal size, together with an accentuation of the intra-crystalline shrinkage. This is presumably due to the smaller freezing range caused by the greater homogeneity of the slowly frozen metal.

Pouring Group 2 Alloys

40. The second group of metals are typified by the aluminum bronzes and the manganese bronzes. No great amount of these metals has been poured by other than conventional methods, but such work as has been done indicates that pouring temperatures for top-gating must be as high or higher than when the normal bottom gates are used. The first, and most important reason, is that if too low temperatures are used, dross entrapment results. Experience seems to show that simple castings should remain liquid for at least a minute per foot of vertical height if the dross formed on pouring is to float out to give good physical properties. Where freezing is more rapid, the condition shown at the left in Fig. 1 may result.

41. It is fortunate that the iron content of these alloys is such that the desirably small crystal size is maintained in spite of the somewhat higher temperatures used. In addition, the non-dendritic type of crystal, found in these alloys, does not produce the same sort of shrinkage as is found in the first class of alloys and, as a result, feeding is ample.

CONCLUSION

42. It has been attempted, in the development of this paper, to show that the use of top or riser gating is a very potent foundry tool that may be applied to many types of castings and to many different metals.

43. The author believes that the logical usage of this type of pouring will solve some foundry problems and, in many cases, will allow the production of sound castings at less cost. The author has given his experiences in the development of a technique and while he knows that better methods will be evolved, he hopes that these methods will offer a beginning point for application that has not already been made.

DISCUSSION

Presiding: W. ROMANOFF, H. Kramer & Co., Chicago, Ill.

MEMBER: What is the ratio between the weight of the casting and the weight of the risers?

MR. HIGGINS: The actual total weight of risers is probably on the order of 15 to 20 per cent. The thickness dimensions are one and a half times the wall thickness—or slightly over.

B. A. MILLER[1]: I heard you mention manganese-bronze top gating. Is that so?

MR. HIGGINS: We poured a small amount of it that way. In certain castings, we have had very good results. We have not tried it in a great many cases, because it does seem rather "heresy" to do it that way.

MR. MILLER: What size castings and what metal cross section could be poured using that method? Say, a bushing 80-in. outside diameter, 70-in. inside diameter and 40-in. deep?

MR. HIGGINS: We have not poured any bushings to the present time. We have poured simple stick castings such as those illustrated in the paper. The largest we have poured is about a 3-in. diameter. We have poured small castings.

MR. MILLER: Did you find in the tests that you were able to increase the casting yield by top-pouring manganese-bronze?

MR. HIGGINS: We made no tests on that particular material.

MR. MILLER: Were you able to use less riser?

MR. HIGGINS: We were able to use much less riser; I would say about 30 per cent of the normal riser.

MR. MILLER: How about the higher nickel alloys—up to 20 per cent nickel plus?

MR. HIGGINS: We poured some castings of a patented high nickel alloy. We try to drop it free to the bottom of the mold so there is a metal cushion into which the metal falls subsequent to first pouring. Gates up to 1-in. in diameter are used so that the castings fill quite rapidly.

B. A. MILLER: Do you notice a decrease of riser on the nickel alloys as well?

MR. HIGGINS: Yes.

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[1] Metallurgist, Cramp Brass & Iron Foundries Co., Philadelphia, Pa.

HAROLD J. ROAST[2]: In our foundry, we have gone very considerably into this top-pouring method with castings from 100 to 6,000 lb. In the 6,000 lb. castings, particularly in those with runners which have 22 or 26 thin veins about 2 ft. high and entering into heavy rings top and bottom, we have found it successful in high shrinkage alloys, but so far we have not tried manganese or aluminum bronze.

My only reason for mentioning that is that some foundrymen are doubtful about trying it. Do not hesitate. If you have a little trouble, keep at it because there is a great deal to be gained by using this top-pouring method. Do not be afraid of churning up the sand. Take precautions suggested by the paper and I feel sure excellent results will be obtained.

MR. MILLER: I would like to ask you if you have tried it on your silicon-bronze.

HAROLD J. ROAST: That is what we used in our 6,000 lb. castings. It worked out very satisfactorily.

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[2] Vice-president, Canadian Bronze Co., Ltd., Montreal, Que., Can.

The Pearlite Interval in Gray Cast Iron

By Alfred Boyles,* Columbus, Ohio.

Abstract

In this paper, the author records the results obtained in a continuation of the studies in which he has been engaged, namely, studies of the fundamental concepts of cast iron. In the section of the projects presented here, the author has studied the pearlite transformation range in both the stable and metastable conditions of the system iron-carbon-silicon. For his work, he chose two irons of the following analysis and properties: (1) Total carbon, 2.93 per cent; silicon, 2.19 per cent; manganese, 0.76 per cent; sulphur, 0.064 per cent; phosphorus, 0.093 per cent; transverse strength, 2540 lb.; deflection, 0.215 in.; tensile strength, 32,500 lb. per sq. in.; Brinell hardness, 190; (2) Total carbon, 3.03 per cent; silicon, 2.34 per cent; manganese, 0.70 per cent; sulphur, 0.063 per cent; phosphorus, 0.097 per cent; transverse strength, 2960 lb.; deflection, 0.415 in.; tensile strength, 41,000 lb. per sq. in.; Brinell hardness, 192. From test bars from these irons, the author selected specimens and heat treated them in such a way as to arrest the transformation at various points throughout the range. He presents micrographs of his study to show that ferrite, austenite and graphite exist in equilibrium within a range of 1450 to 1550°F. for the irons studied. Other results of his studies on the same irons showed that (1) the rate of formation of both ferrite and pearlite was accelerated at subcritical temperatures, (2) under identical conditions of heat treatment, an iron containing fine graphite flakes showed more ferrite than one containing large graphite flakes, (3) initial temperature of heating had an effect on the rate of formation of ferrite, (4) in small castings, examined by quenching at various cooling stages in the mold, ferrite began to form along the graphite flakes prior to the formation of pearlite, continued to develop during the transformation period and that no additional ferrite ap-

* Research Metallurgist, Battelle Memorial Institute.

Note: This paper was presented at a Gray Iron Session at the 44th A.F.A. Convention, May 10, 1940, Chicago, Ill.

peared after transformation was complete and (5) that silicon not only promotes graphitization but also provides a mechanism for the formation of free ferrite.

1. In much of the literature, cast iron has been described as an alloy of iron and carbon in which the element silicon promotes graphitization. From this point of view, the matrix is simply a steel in which graphite flakes are imbedded and its behavior during cooling is essentially the same as that of a steel of hypereutectoid composition which rejects carbide down to the eutectoid temperature and then transforms into pearlite. It is recognized that cast iron may reject graphite instead of carbide during the proeutectoid interval, such graphite being added onto the flakes already present. The behavior is then interpreted in the light of the iron-carbon diagram according to either the stable or metastable system. Ordinarily, both stable and metastable conditions are exhibited in a single casting, the formation of proeutectoid graphite representing the iron-graphite system and the formation of pearlite the iron-cementite system. In addition, there is often a considerable amount of free ferrite present along the graphite flakes and this, in turn, is said to result either from the graphitization of pearlite or from the breakdown of a hypothetical "graphite eutectoid," both cases requiring a migration of carbon to the graphite flakes.

2. The manner in which the ferrite forms is a point of considerable interest. Castings often show an amount of ferrite out of all proportion to what would be expected either from the graphitization of pearlite or the divorce of a "graphite eutectoid." The present work is concerned with the rate of formation of ferrite during transformation and attempts an explanation in terms of the iron, carbon, silicon equilibrium diagram.

3. Fig. 1 shows three sections through the ternary diagram of iron, carbon, silicon system according to Greiner, Marsh, and Stoughton[1]. This diagram represents the metastable equilibrium between iron and iron carbide, no account being taken of graphitization. In the absence of silicon (Fig. 1—Bottom), the pearlite reaction is indicated by a single line, *i.e.*, the phases ferrite, austenite and carbide are in equilibrium at a single temperature. With silicon present (Fig. 1—Center and Top), these phases are in equilibrium over a range of temperature, the so-called pearlite interval, represented by the region marked $\alpha + \gamma + Ca$.

[1] Superior numbers refer to references at the end of this paper.

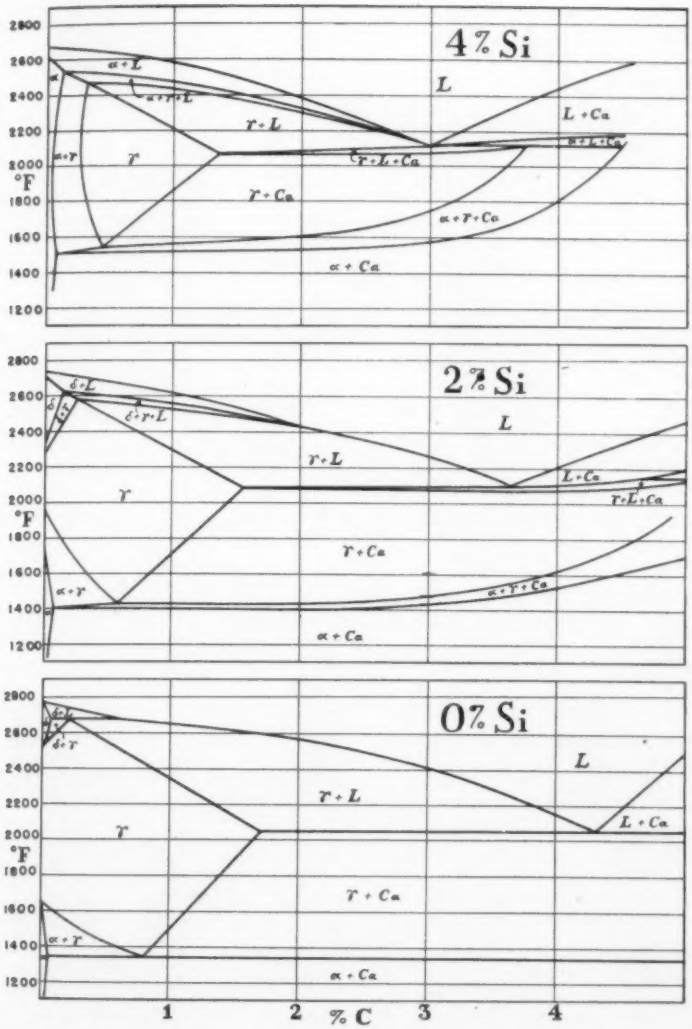

FIG. 1—THREE SECTIONS THROUGH THE TERNARY DIAGRAM OF THE IRON, CARBON, SILICON SYSTEM (ACCORDING TO GREINER, MARSH AND STOUGHTON).

4. In attempting to apply these diagrams to gray cast iron, some account must be taken of graphitization because any cooling rate sufficient to approach equilibrium must involve a breakdown of some of the carbide. For this reason, the first thing undertaken was a determination of the temperature range in which ferrite,

austenite, and graphite exist in equilibrium in ordinary gray iron containing about 3 per cent carbon and 2 per cent silicon.

Materials Used

5. As experimental material, two cast irons were selected having widely different types of graphite structure, the analyses and physical properties of which are shown in Table 1.

Table 1

ANALYSIS AND PHYSICAL PROPERTIES OF CAST IRONS USED TO DETERMINE RANGE IN WHICH FERRITE, AUSTENITE AND GRAPHITE EXIST IN EQUILIBRIUM.

	Iron No. 46	*Iron No. 149*
Total Carbon, per cent	2.93	3.03
Silicon, per cent	2.19	2.34
Manganese, per cent	0.76	0.70
Sulphur, per cent	0.064	0.063
Phosphorus, per cent	0.093	0.097
Transverse Strength, lb	2540	2960
Deflection, in	0.215	0.415
Tensile Strength, lb. per sq. in	32,500	41,000
Brinell Hardness Number	190	192

6. Iron No. 46 was made from pig and steel melted in an induction furnace, superheated to 3050°F. and poured at 2550°F. without ladle additions. The transverse test bars have a highly modified structure consisting of very fine graphite flakes in a matrix of pearlite with scattered areas of free ferrite.

7. Iron No. 149 was made from pig and steel melted in an induction furnace, heated to 2550°F. and poured at 2550°F. after receiving a ladle addition of 0.5 per cent silicon (as 75 per cent ferrosilicon). Its structure is completely normal, consisting of medium sized graphite flakes in a matrix of pearlite with some free ferrite along the graphite flakes.

8. Specimens, weighing 40 grams, were cut from the transverse test bars and subjected to the treatments listed in Table 2. Annealing was conducted in a vertical-tube globar furnace, automatically controlled and evacuated to a pressure of about one millimeter of mercury. The specimens were suspended from wires so that, after annealing was complete, they could either be dropped

Table 2

ANNEALING TREATMENTS OF GRAY IRON

Specimen No.	*Iron No.*	*Treatment*
292	149	Held 96 hours at 1600°F. and Quenched
294	149	Held 96 hours at 1600°F. and Cooled in air
291	46	Held 96 hours at 1600°F. and Quenched
293	46	Held 96 hours at 1600°F. and Cooled in air
305	149	Held 96 hours at 1550°F. and Quenched
307	149	Held 96 hours at 1550°F. and Cooled in air
304	46	Held 96 hours at 1550°F. and Quenched
306	46	Held 96 hours at 1550°F. and Cooled in air
301	149	Held 96 hours at 1500°F. and Quenched
303	149	Held 96 hours at 1500°F. and Cooled in air
300	46	Held 96 hours at 1500°F. and Quenched
302	46	Held 96 hours at 1500°F. and Cooled in air
309	149	Held 96 hours at 1450°F. and Quenched
311	149	Held 96 hours at 1450°F. and Cooled in air
308	46	Held 96 hours at 1450°F. and Quenched
310	46	Held 96 hours at 1450°F. and Cooled in air
313	149	Held 96 hours at 1400°F. and Quenched
315	149	Held 96 hours at 1400°F. and Cooled in air
312	46	Held 96 hours at 1400°F. and Quenched
314	46	Held 96 hours at 1400°F. and Cooled in air

directly into water or pulled out to cool in air. Each specimen was sectioned through the center for microexamination.

Structure of the Specimens

9. Annealing for 96 hr. at 1600°F. and water quenching gave a fully martensitic structure (Figs. 2—Left and 3—Left). The com-

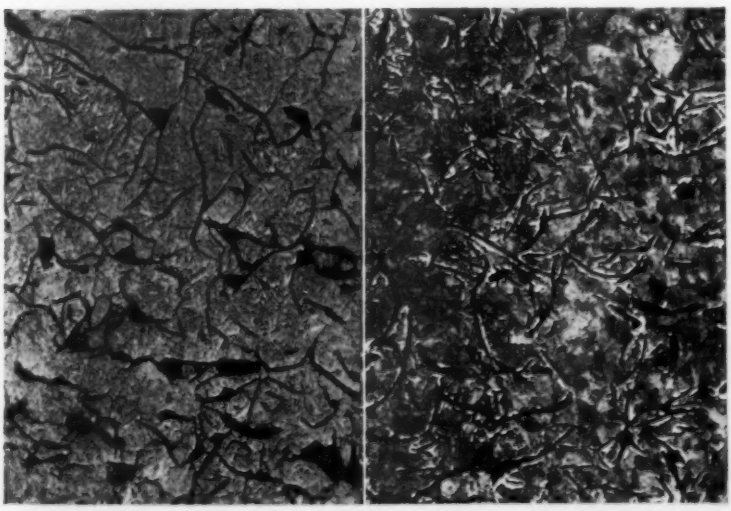

FIG. 2—LEFT—IRON NO. 149 HELD 96 HR. AT 1600°F. AND QUENCHED. ETCHED, X100. RIGHT—IRON NO. 149 HELD 96 HR. AT 1600°F. AND COOLED IN AIR. ETCHED, X100.

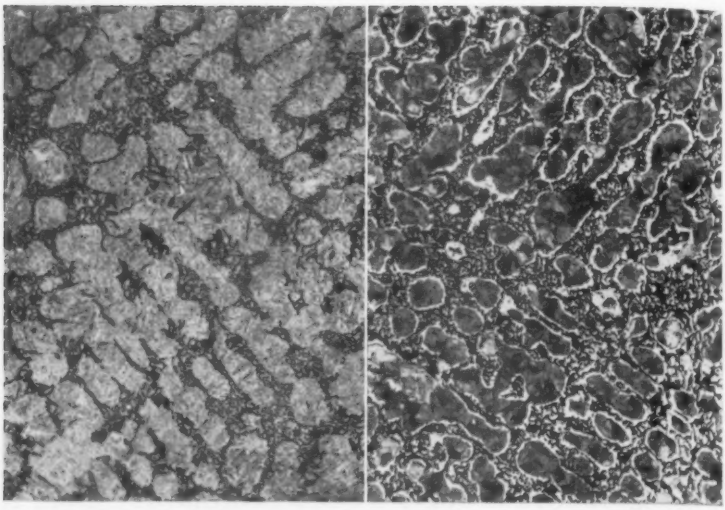

FIG. 3—LEFT—IRON NO. 46 HELD 96 HR. AT 1600°F. AND QUENCHED. ETCHED, X100. RIGHT—IRON NO. 46 HELD 96 HR. AT 1600°F. AND COOLED IN AIR. ETCHED, X100.

panion pieces, air cooled after annealing, are shown in Figs. 2—Right and 3—Right. The normal iron consists of pearlite with some free ferrite along the graphite flakes. In the modified iron, the fine graphite is completely surrounded by ferrite, pearlite occurring along the midribs of the primary dendrites.

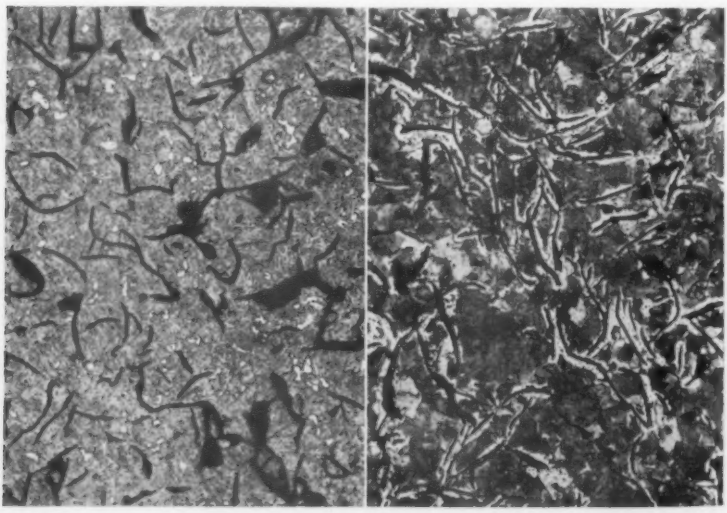

FIG. 4—LEFT—IRON NO. 149 HELD 96 HR. AT 1550°F. AND QUENCHED. ETCHED, X100. RIGHT—IRON NO. 149 HELD 96 HR. AT 1550°F. AND COOLED IN AIR. ETCHED, X100.

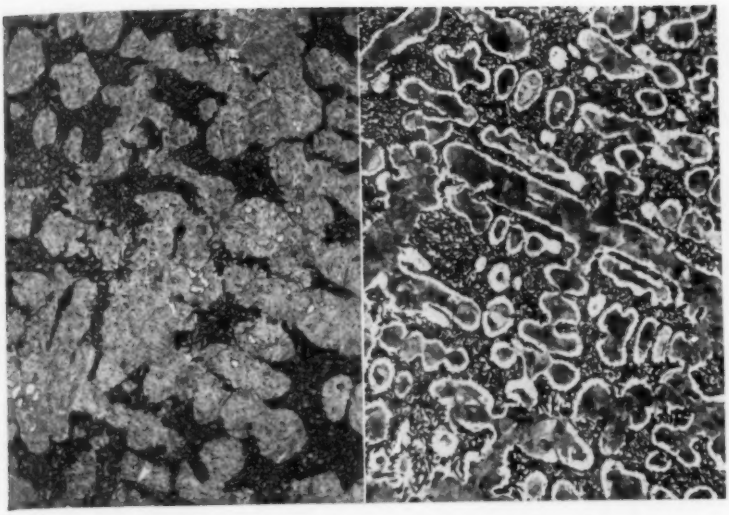

FIG. 5—LEFT—IRON NO. 46 HELD 96 HR. AT 1550°F. AND QUENCHED. ETCHED, X100. RIGHT—IRON NO. 46 HELD 96 HR. AT 1550°F. AND COOLED IN AIR. ETCHED, X100.

10. Annealing for 96 hr. at 1550°F. and quenching, gave the structures shown in Figs. 4—Left and 5—Left. In both irons, a small amount of ferrite occurs as tiny particles located mainly in the spaces between the graphite flakes, the rest of the matrix being

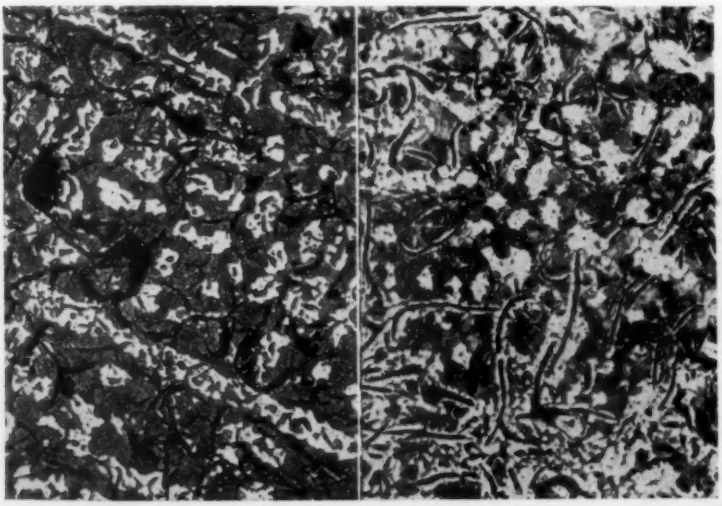

FIG. 6—LEFT—IRON NO. 149 HELD 96 HR. AT 1500°F. AND QUENCHED. ETCHED, X100. RIGHT—IRON NO. 149 HELD 96 HR. AT 1500°F. AND COOLED IN AIR. ETCHED, X100.

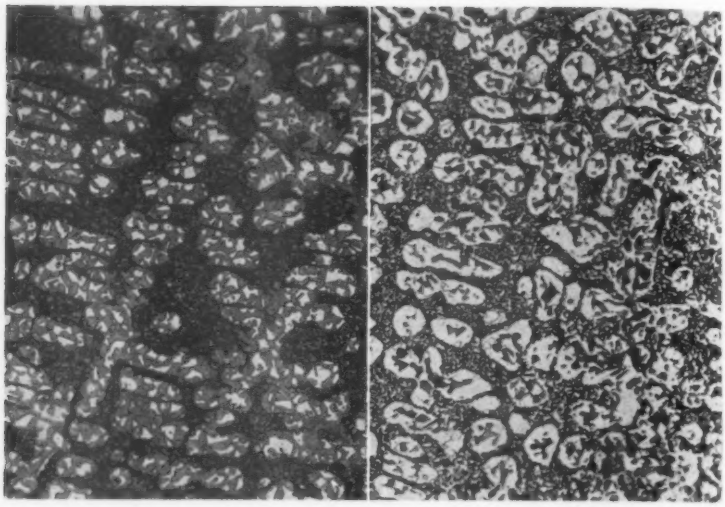

FIG. 7—LEFT—IRON NO. 46 HELD 56 HR. AT 1500°F. AND QUENCHED. ETCHED, X100. RIGHT—IRON NO. 46 HELD 96 HR. AT 1500°F. AND COOLED IN AIR. ETCHED, X100.

martensite. Air cooling after annealing at 1550°F., gave the structures shown in Figs. 4—Right and 5—Right. The normal iron consists of pearlite with ferrite adjacent to the graphite flakes, while the modified iron shows pearlite along the midribs of the dendrites, the remainder of the matrix being ferrite.

11. Annealing for 96 hr. at 1500°F. and quenching, gave the structures shown in Figs. 6—Left and 7—Left. Ferrite occurs as angular masses quite definitely arranged along the midribs of the primary dendrites, the remainder of the matrix being martensite. The specimens cooled in air show a composite ferrite structure consisting of the ferrite already present after annealing plus that formed during cooling (Figs. 6—Right and 7—Right). Pearlite occurs as irregular patches scattered through the matrix, the modified iron containing a much smaller amount than the normal iron.

12. Annealing for 96 hr. at 1450°F. and quenching resulted in the structures shown in Fig. 8—Left. Traces of martensite occur as tiny patches adjacent to the graphite flakes (Fig. 9—Left). The air cooled specimens were identical to Fig. 8—Right, except for the presence of minute patches of pearlite which represent the transformation product of the austenite remaining after the annealing treatment (Fig. 9—Right).

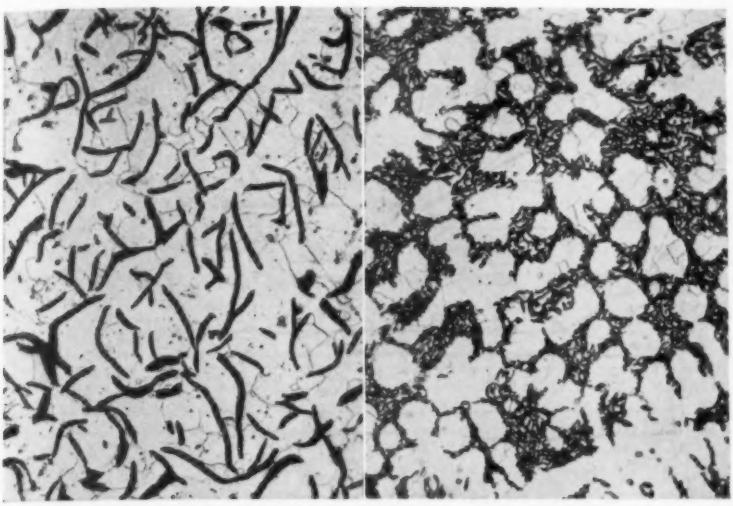

FIG. 8—LEFT—IRON NO. 149 HELD 96 HR. AT 1450°F. AND QUENCHED. ETCHED, X100. RIGHT—IRON NO. 46 HELD 96 HR. AT 1450°F. AND QUENCHED. ETCHED, X100.

13. Specimens annealed for 96 hr. at 1400°F. were identical to those treated at 1450°F., except that no trace of martensite could be found in the quenched samples. The air cooled specimens were completely ferritic with no trace of pearlite.

FIG. 9—LEFT—IRON NO. 46 HELD 96 HR. AT 1450°F. AND QUENCHED. THE ARROWS POINT TO MARTENSITE. ETCHED, X1000. RIGHT—IRON NO. 46 HELD 96 HR. AT 1450°F. AND COOLED IN AIR. THE ARROW POINTS TO PEARLITE. ETCHED, X1000.

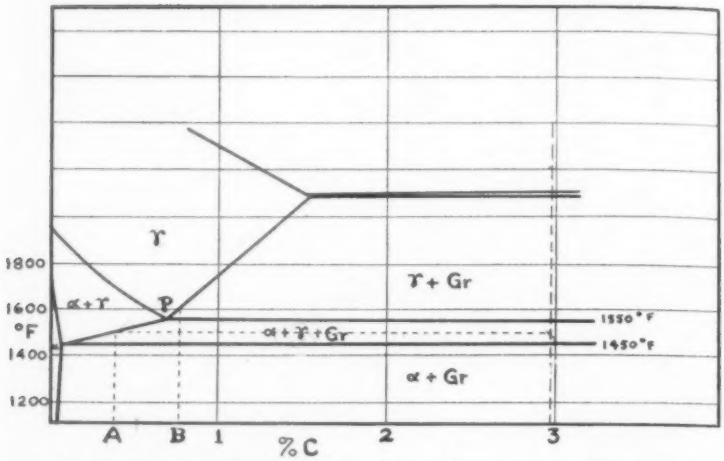

FIG. 10—SCHEMATIC DIAGRAM OF THE STABLE EQUILIBRIUM IN GRAY IRON.

Redrawing of Stable Equilibrium Diagram

14. These specimens represent an approach to a condition of stable equilibrium for the particular temperatures employed. Assuming that no graphitization occurred during the quench, the results indicate that ferrite, austenite, and graphite are in equilibrium over a range of temperature extending from 1450° to 1550°F. This temperature range corresponds to the pearlite interval in the metastable system (the region $\alpha + \gamma + Ca$ in Fig. 1) and represents the interval during which the transformation of austenite would occur under conditions of stable equilibrium.

15. To make this clearer, the lower part of the metastable equilibrium diagram may be redrawn as shown in Fig. 10 and an imaginary case of transformation according to the stable system followed through. Fig. 10 is simply a copy of the 2 per cent silicon section of Fig. 1 with the region $\alpha + \gamma + Ca$ changed to $\alpha + \gamma + Gr$ and widened to embrace the temperature range between 1450° and 1550°F. In the other regions of the diagram the notation "*Ca*" (carbide) has been changed to "*Gr*" (graphite).

Interpretation of Redrawn Stable Diagram

16. Taking the case of an iron of 3 per cent total carbon, we may trace out the changes in the matrix during cooling, assuming that ample time is allowed at successively lower temperatures for the alloy to reach a condition of stable equilibrium. At 1600°F., the matrix consists of austenite holding in solid solution a percen-

tage of carbon equal to point "*B*" on the diagram. Upon cooling, the austenite deposits carbon on the existing graphite flakes until the matrix reaches eutectoid composition at about 1550°F. At this temperature, it enters the region $\alpha + \gamma + Gr$ and part of the austenite begins to break down into ferrite and graphite. Since a graphite eutectoid structure was not observed, it may be assumed that the graphite precipitated is added on to the existing flakes. Figs. 4—Left and 5—Left, held for 96 hr. at 1550°F., represent an early stage in the ferrite formation. The matrix here is almost all austenite of eutectoid composition corresponding to point "*P*" in the diagram.

17. Upon cooling further, the decomposition of the austenite continues, ferrite being deposited and carbon diffusing to the graphite flakes. Figs. 6—Left and 7—Left show this stage of the transformation. According to the diagram, the matrix at 1500°F. should consist of austenite of composition "*P*" plus ferrite (which holds in solution a small amount of carbon), the average carbon content of the matrix being represented by point "*A*". At successively lower temperatures, the amount of ferrite increases while the amount of austenite diminishes until transformation is finally complete at about 1450°F. Fig. 8 represents the condition of the matrix near the end of transformation. The last austenite to transform is that immediately adjacent to the graphite flakes (Fig. 9—Left). At temperatures below 1450°F., there is no further change in the matrix, which consists of polyhedral grains of ferrite.

18. The mechanism just outlined assumes a condition of stable equilibrium to be reached at all temperatures. With this as a basis, an attempt may be made to describe the transformation as it occurs with faster rates of cooling. The matrix in Fig. 4—Left is very near eutectoid composition, only a few traces of ferrite being present. Upon cooling such a specimen in air, the structure shown in Fig. 4—Right resulted. A certain amount of ferrite has formed along the graphite flakes, the rest of the matrix being pearlite. The formation of the ferrite layers may be explained as follows.

Formation of Ferrite Layers

19. As soon as the metal entered the region $\alpha + \gamma + Gr$, a diffusion of carbon to the graphite flakes started. A layer of ferrite formed along the graphite-austenite interface while graphite was deposited onto the flake surface. This action continued until cooling brought it to a stop and the remaining austenite, still of

eutectoid composition, transformed into pearlite. Specimens cooled in air after prolonged annealing in the region $\alpha + \gamma + Gr$ show a composite ferrite structure consisting of the ferrite already present in the annealed specimen plus that formed during cooling. Figs. 6—Right and 7—Right illustrate this condition. It will be noted by comparing Fig. 6—Right with Fig. 4—Right that the amount of ferrite formed along the graphite flakes is practically the same in both cases. The additional ferrite in Fig. 6—Right is simply that already present when cooling started.

20. A comparison of the quenched specimens with those cooled in air also indicates that the rate of formation of ferrite must vary at different temperatures. In Fig. 5—Left, for example, there is only a trace of ferrite present after holding 96 hr. at 1550°F. The same material, air cooled, shows a considerable amount (Fig. 5—Right). Fig. 7—Left shows less ferrite formed by annealing 96 hr. at 1500°F. than was formed during air cooling from the same temperature (Fig. 7—Right). Regardless of how the ferrite formed, its rate of formation must have been accelerated as the metal cooled below 1500°F. To study this question the following experiment was carried out.

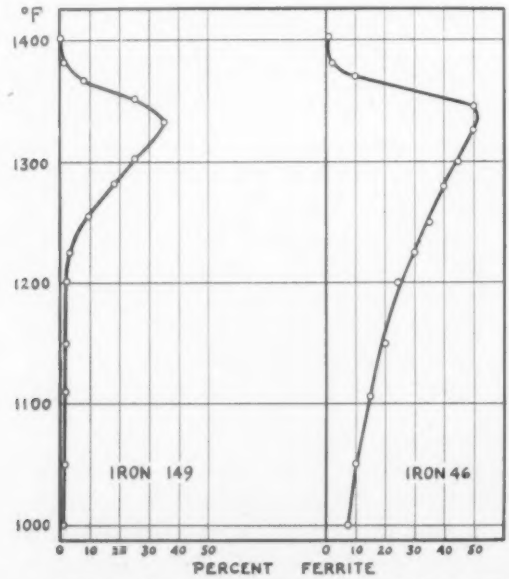

FIG. 11—AMOUNT OF FERRITE FORMED IN 5 MIN. AT VARIOUS TEMPERATURES. SPECIMENS QUENCHED FROM 1600°F. INTO A LEAD BATH.

Table 3

SPECIMENS HELD FIVE MINUTES IN LEAD BATH AND QUENCHED IN WATER.

Specimen No.	*Iron No.*	*Lead Temp., °F.*	*Structure of Matrix*
321	149	1556	Martensite
322	149	1505	Martensite
323	149	1454	Martensite
324	149	1400	Martensite + Trace Ferrite
329	149	1383	Martensite + Ferrite
327	149	1367	Martensite + Ferrite + Trace Pearlite
325	149	1352	Martensite + Ferrite + Pearlite
331	149	1330	Trace Martensite + Ferrite + Pearlite
333	149	1300	Ferrite + Pearlite
335	149	1280	Ferrite + Pearlite
343	149	1252	Ferrite + Pearlite
341	149	1225	Trace Ferrite + Pearlite
339	149	1203	Trace Ferrite + Pearlite
337	149	1150	Trace Ferrite + Pearlite
349	149	1110	Trace Ferrite + Pearlite
347	149	1050	Trace Ferrite + Pearlite
345	149	1000	Trace Ferrite + Pearlite
320	46	1556	Martensite
319	46	1506	Martensite
318	46	1454	Martensite
317	46	1405	Martensite + Trace Ferrite
328	46	1382	Martensite + Ferrite
326	46	1372	Martensite + Ferrite + Trace Pearlite
316	46	1345	Martensite + Ferrite + Pearlite
330	46	1326	Trace Martensite + Ferrite + Pearlite
332	46	1300	Ferrite + Pearlite
334	46	1278	Ferrite + Pearlite
342	46	1250	Ferrite + Pearlite
340	46	1225	Ferrite + Pearlite
338	46	1200	Ferrite + Pearlite
336	46	1150	Ferrite + Pearlite
348	46	1107	Ferrite + Pearlite
346	46	1050	Ferrite + Pearlite
344	46	1000	Ferrite + Pearlite

AMOUNT OF FERRITE FORMED AT VARIOUS TEMPERATURES

21. Specimens weighing 40 grams were heated to 1600°F. and held for one hour. They were then quenched into a lead bath, standing at constant temperature and held for 5 min. and finally quenched into water. Table 3 lists the specimens examined and the range of lead temperatures covered as well as the structural constituents present. To summarize the data graphically, the percentage of ferrite was estimated by the method given in the appendix and the results plotted in Fig. 11.

Structure of the Specimens

22. Specimens treated at 1454°F. and above were fully martensitic, indicating that no transformation occurred during the time

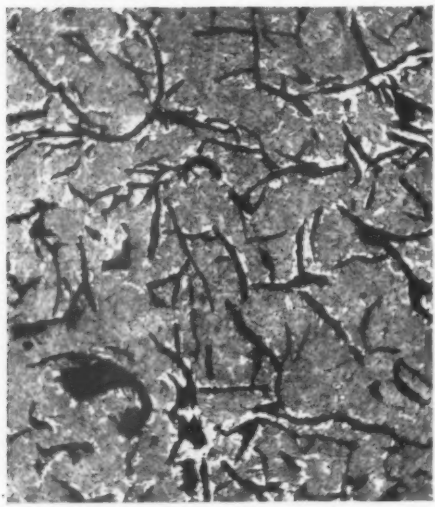

FIG. 12—IRON NO. 149 QUENCHED TO 1367°F. AND HELD 5 MIN. MARTENSITE + FERRITE + TRACE OF PEARLITE. ETCHED, X100.

they were held in the lead bath. At 1400°F., the first traces of free ferrite appear, the rest of the matrix being martensite. Ferrite increased in amount at lower temperatures reaching a maximum

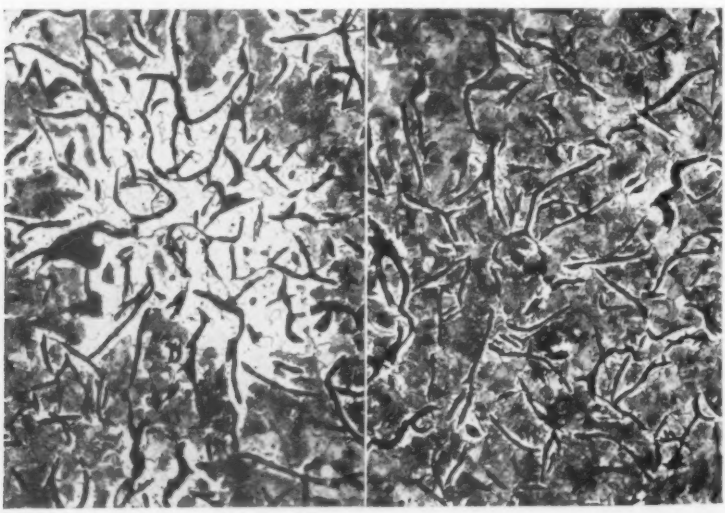

FIG. 13—LEFT—IRON NO. 149 QUENCHED TO 1330°F. AND HELD 5 MIN. FERRITE + PEARLITE + TRACE OF MARTENSITE. ETCHED, X100. RIGHT—IRON NO. 149 QUENCHED TO 1280°F. AND HELD 5 MIN. FERRITE + PEARLITE. ETCHED, X100.

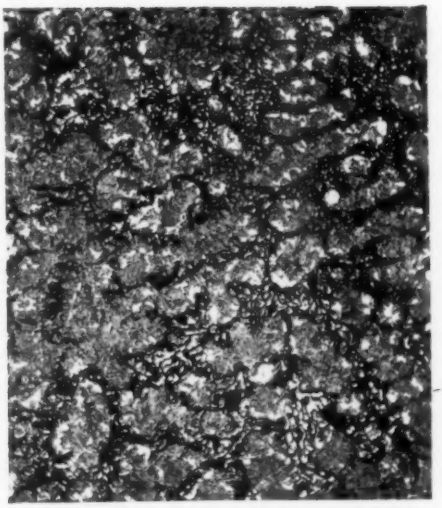

FIG. 14—IRON NO. 46 QUENCHED TO 1372°F. AND HELD 5 MIN. MARTENSITE + FERRITE + TRACE OF PEARLITE. ETCHED, x100.

around 1330°F. (Figs. 13—Left and 15—Left). At lower temperatures, the amount of ferrite decreased (Figs. 13—Right and 15—Right) reaching a minimum at 1000°F. In all cases, the modified iron showed more ferrite than the normal iron.

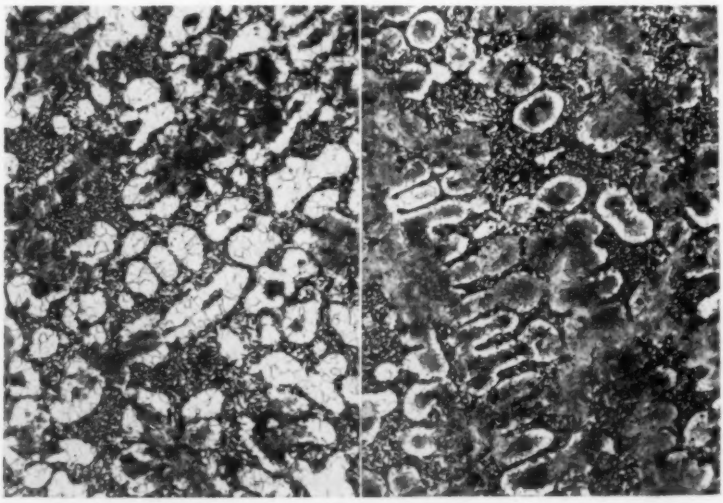

FIG. 15—LEFT—IRON NO. 46 QUENCHED TO 1326°F. AND HELD 5 MIN. FERRITE + PEARLITE + TRACE OF MARTENSITE. ETCHED, x100. RIGHT—IRON NO. 46 QUENCHED TO 1278°F. AND HELD 5 MIN. FERRITE + PEARLITE. ETCHED, x100.

23. The first traces of pearlite appeared in the specimens treated at 1367°F. and 1372°F. (Figs. 12 and 14). This pearlite occurred as tiny patches detectable only at high magnification. At 1326 and 1330°F., the entire matrix consisted of pearlite and ferrite with the exception of a few traces of martensite. At all temperatures below this, no trace of martensite was found, the matrix having transformed completely in the lead bath. The pearlite lamellae became finer at successively lower temperatures.

24. These results show that the rate of formation of ferrite is accelerated at temperatures below the stable transformation range (the region $\alpha + \gamma + Gr$ in Fig. 10). They do not indicate that such ferrite resulted from a transformation according to the stable system because all the ferrite in Figs. 12 to 15 may have resulted from the graphitization of previously existing pearlite. Although ferrite appeared first in the specimens treated at 1400°F., it may still have formed by the breakdown of pearlite during the time the specimens were in the lead bath.

25. These specimens represent only one time of holding at any given temperature and, for this particular condition, the structures indicate the range of temperature most favorable to the formation of ferrite. They do not give any information as to the time required for the transformation to complete itself nor the relative rates of formation of pearlite and ferrite during transformation.

26. To determine this, specimens must be held for various lengths of time at each particular temperature. The series covered a wide range of temperature in rather small steps and made possible the selection of certain temperatures as being most suitable for further study.

TRANSFORMATION RATE AT CONSTANT SUBCRITICAL TEMPERATURES

27. Pieces of uniform size and shape were cut from the same transverse test bars for this work. Each specimen was fastened to the end of a stiff chromel wire and placed in the tube furnace standing at 1600°F. After holding one hour, each specimen was quenched into a lead bath standing at the desired temperature, held for a definite time and finally quenched in water. The time required to transfer from the heating furnace to the lead pot was about one second. By means of a fixture, each specimen was immersed to the same depth in the lead and held immediately adjacent the hot junction of the thermocouple used to measure the lead temperature.

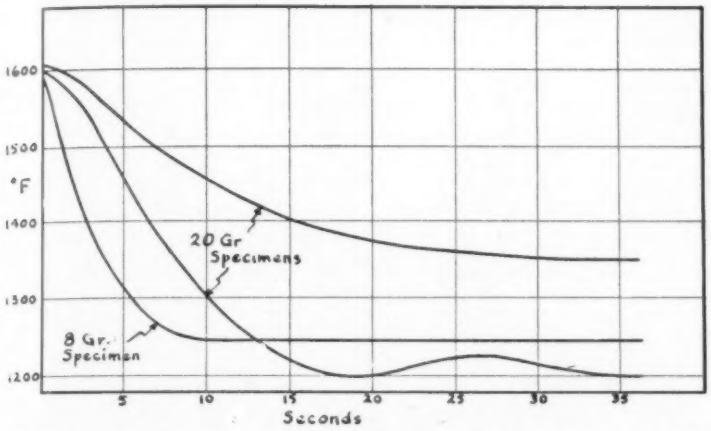

FIG. 16—COOLING CURVES OF SPECIMENS QUENCHED IN LEAD. PLOTTED FROM STRING GALVANOMETER RECORDS.

28. To determine the time required for the specimens to cool from 1600°F. to the temperature of the lead bath, a number of records were made with a string galvanometer. A dummy specimen was drilled to take a platinum platinum-rhodium thermocouple with the hot junction at the center of mass and in direct contact with the iron. This was connected to the string circuit of an electrocardiograph having a range of 6 millivolts. By placing a potentiometer in series, the E.M.F. developed by the thermocouple at 1600°F. was balanced out so that no deflection was shown by the string. Upon quenching the dummy specimen into lead in the usual way, a photographic record of the cooling curve was taken by the cardiograph camera. Since the temperatures of the beginning and end of the cardiograph curve were known, the plotting of the records in terms of degrees F. was easily carried out.

29. Fig. 16 shows several of the cooling curves for specimens of two sizes. The recalescence shown by the 20-gram specimen quenched to 1200°F. should be noted as it checks very nicely with the progress of transformation observed in the series of specimens treated at 1200°F.

Structure of the Specimens

30. A section for microexamination was cut through the center of each specimen and the percentages of ferrite and pearlite by area were estimated and plotted on summary charts (see appendix). In Figs. 17 and 18, the height of the vertical bars repre-

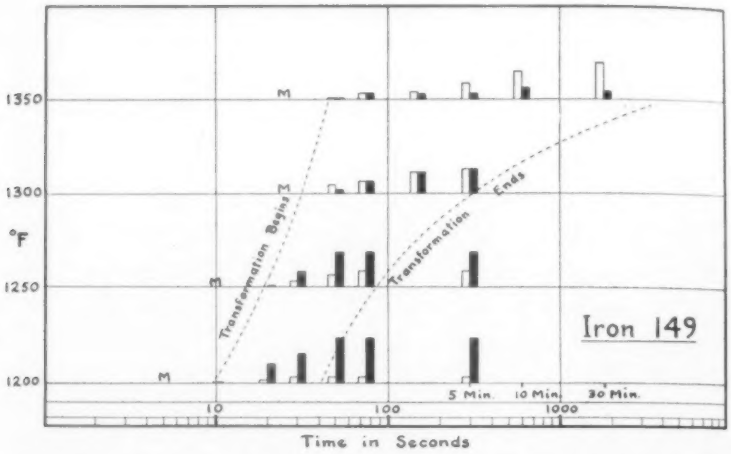

FIG. 17—SUMMARY CHART OF IRON NO. 149. BLACK BARS REPRESENT PEARLITE AND WHITE BARS FERRITE. SPECIMENS QUENCHED FROM 1600°F.

sents the amount of ferrite and pearlite present in each specimen, white bars indicating ferrite and black bars pearlite. The letter "*M*" indicates a fully martensitic structure. Time of holding in the lead bath is plotted as seconds on a logarithmic scale. Dotted lines indicate approximately the beginning and end of transformation at each temperature level. Micrographs of the series transformed at 1300°F. will suffice to illustrate the type of structures observed.

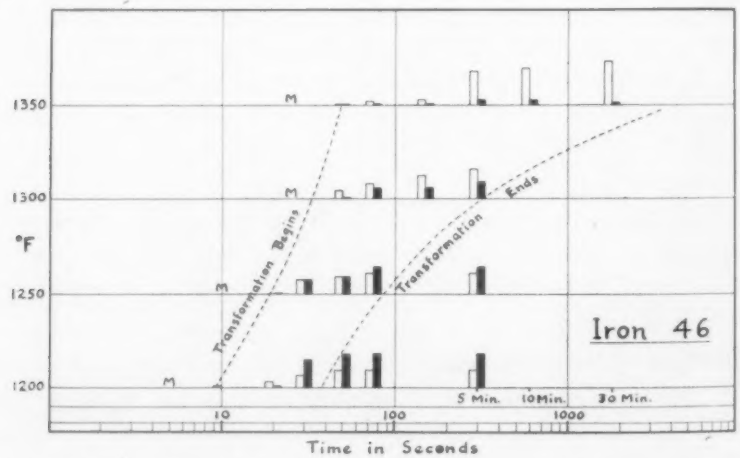

FIG. 18—SUMMARY CHART OF IRON NO. 46. BLACK BARS REPRESENT PEARLITE AND WHITE BARS FERRITE. SPECIMENS QUENCHED FROM 1600°F.

Series Treated at 1350°F. (20-Gram Specimens)

31. Transformation started after 50 sec. with traces of ferrite and pearlite in both irons. After 75 sec., about 7 per cent ferrite had formed in both irons and the pearlite formation was under way at the same time. The amount of ferrite increased gradually, reaching 70 per cent after 30 min. The pearlite increased very slowly

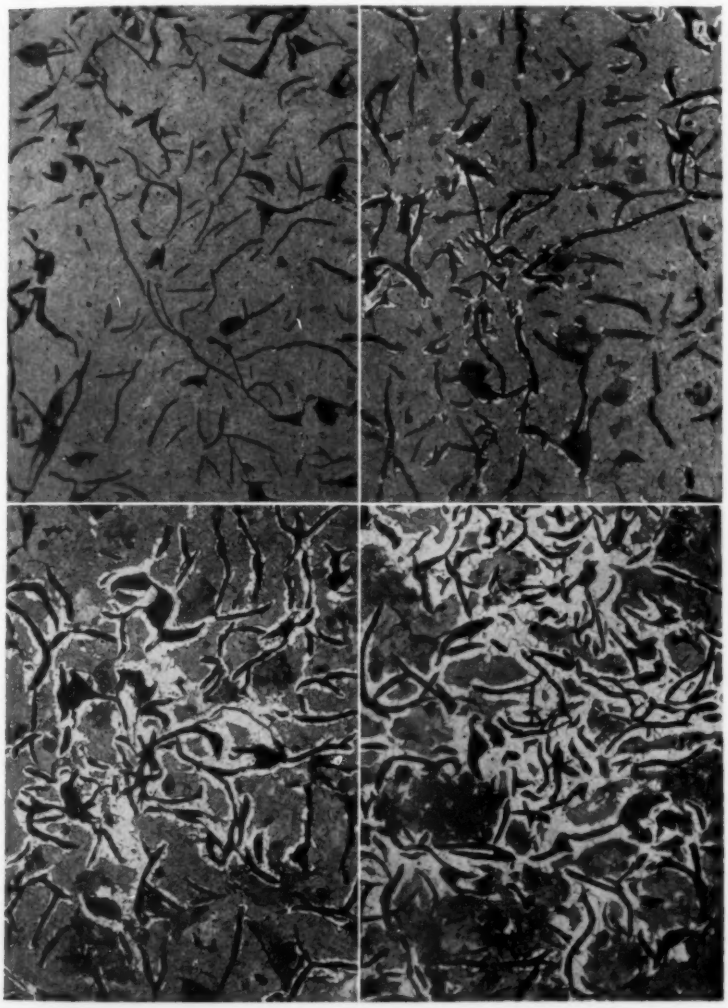

FIG. 19—IRON NO. 149—QUENCHED FROM 1600°F. TO 1300°F. AND HELD FOR TIME INDICATED. TOP LEFT—HELD 25 SEC. TOP RIGHT—HELD 50 SEC. BOTTOM LEFT—HELD 75 SEC. BOTTOM RIGHT—HELD 2½ MIN. ALL ETCHED, X100.

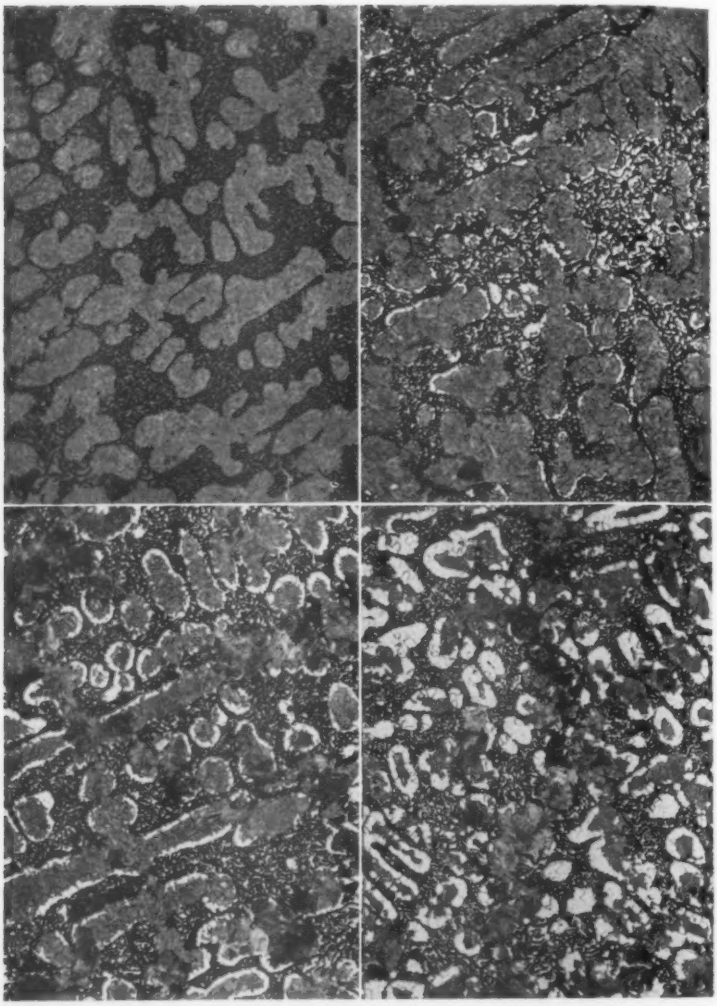

FIG. 20—IRON NO. 46—QUENCHED FROM 1600°F. TO 1300°F. AND HELD FOR TIME INDICATED. TOP LEFT—HELD 25 SEC. TOP RIGHT—HELD 50 SEC. BOTTOM LEFT—HELD 75 SEC. BOTTOM RIGHT—HELD 2½ MIN. ALL ETCHED, x100.

up to 20 per cent in 10 min. in the normal iron. In the specimens held for 30 min., there was distinct evidence that some of the pearlite had graphitized and the percentage decreased in both irons. The partial spheroidization of the pearlite lamellae was clearly evident in these specimens. In both irons, about 5 per cent of the austenite remained untransformed after 30 min.

Series at 1300°F. (20-Gram Specimens)

32. The structures are shown in Figs. 19 and 20. Transformation started some time after 25 sec. After 50 sec. 10 per cent ferrite had formed in the normal iron and 15 per cent in the modified iron (Fig. 19—Top Right and 20—Top Right). The pearlite formation was also under way after 50 sec. Both ferrite and pearlite continued to form at equal rates in the normal iron. (Fig. 17). In the modified iron, ferrite formed more rapidly than pearlite (Fig. 18). After 2½ min., about 20 per cent of the austenite remained and after 5 min. transformation was complete.

Series at 1250°F. (8-Gram Specimens)

33. The size of the specimens was reduced to speed up the rate of cooling from 1600°F. to 1250°F. Specimens held for 11 sec. were fully martensitic. Transformation started after 20 sec. with the formation of a trace of pearlite. After 30 sec., 10 per cent ferrite formed in the normal iron and 25 per cent in the modified iron. In both irons, 25 per cent pearlite formed after 30 sec. After 50 sec., only 5 per cent of the austenite remained untransformed in the normal iron while 20 per cent remained in the modified iron. Transformation was complete in 75 sec. in both irons.

34. It should be noted that the structure after 75 sec. was identical to that obtained by holding 5 min. Ferrite formed rapidly during transformation and did not increase appreciably after transformation was complete.

Series at 1200°F. (20-Gram Specimens)

35. A trace of ferrite appeared after 10 sec. in the modified iron (Fig. 21). According to Fig. 16, this specimen never reached 1200°F. but actually fell to about 1300°F. prior to the water quench. The specimens held for 20 sec. reached 1200°F. and in these, transformation was well under way. In the normal iron, 5 per cent ferrite and 30 per cent pearlite formed while 10 per cent and a trace of pearlite formed in the modified iron. After 30 sec., the pearlite increased to 40 per cent in the modified iron and to 45 per cent in the normal iron. In 50 sec., transformation was complete.

36. It should be noted that the structure after 50 sec. was identical to that obtained by holding 5 min. Here again the ferrite formed rapidly during transformation and did not increase after transformation was complete.

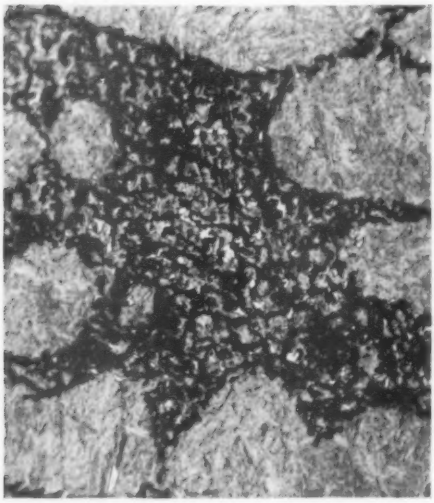

FIG. 21—IRON NO. 46 QUENCHED FROM 1600°F. TO 1200°F. AND HELD 10 SEC. FERRITE IS FORMING BETWEEN THE FINE GRAPHITE FLAKES. ETCHED, X500.

Second Series at 1200°F. (8-Gram Specimens)

37. After making string galvanometer records of the cooling rate down to 1200°F., it was thought advisable to repeat the whole

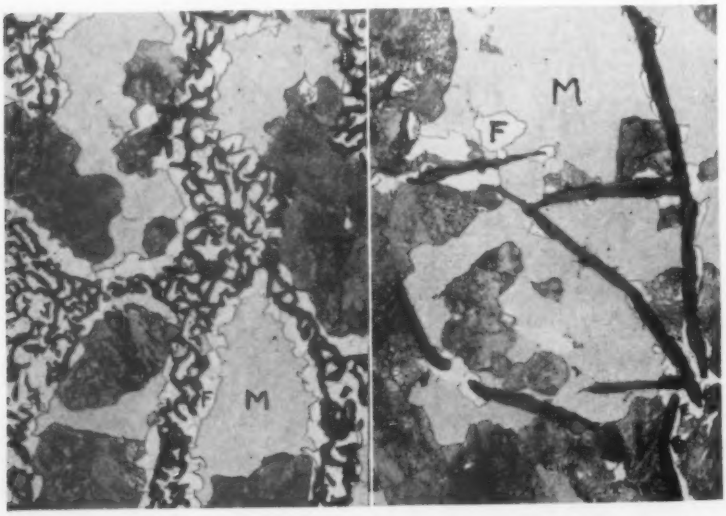

FIG. 22—LEFT—IRON NO. 46 QUENCHED FROM 1600°F. TO 1250°F. AND HELD 30 SEC. MARTENSITE + FERRITE + PEARLITE. ETCHED, X500. RIGHT—IRON NO. 149 QUENCHED FROM 1600°F. TO 1250°F. AND HELD 30 SEC. MARTENSITE + FERRITE + PEARLITE. ETCHED, X500.

series with 8-gram specimens. The structures were in excellent agreement with those obtained on 20-gram specimens and will not be described in detail. Traces of ferrite appeared again in the specimen held for 10 sec. in the lead bath. According to Fig. 16, this specimen had just reached 1200°F. at the time it was water quenched. The formation of ferrite such as that shown in Fig. 21 is evidently very rapid.

38. The following points may be noted in the summary charts (Figs. 17 and 18):

(1) The rate of transformation is accelerated at subcritical temperatures whether expressed in terms of ferrite or pearlite.

(2) At 1350°F. and 1300°F., the rate of ferrite formation exceeds the rate of pearlite formation.

(3) At 1250°F. and 1200°F., the rate of pearlite formation exceeds the rate of ferrite formation.

(4) In all cases, the rate of formation of ferrite is greater in the modified iron.

39. Ferrite started to form along the surface of the graphite flakes in both types of iron (Fig. 22). It did not develop as a continuous layer but grew as small polyhedral grains, the boundaries of which may be seen in Fig. 22—Left. It is highly improbable that the small areas of ferrite in Fig. 22—Right resulted from a breakdown of previously existing pearlite. Some of them are surrounded by martensite which, at the moment of water quenching, was untransformed austenite. There is no trace of any intermediate structure between the martensite and ferrite.

40. At the same time, areas of pearlite may be seen in contact with the graphite flakes with no trace of ferrite between them. Such pearlite is still present after holding 5 min. at 1250°F. The speed with which the ferrite forms is also evidence against the idea that it resulted from a breakdown of pearlite. This is well illustrated in Fig. 21. It seems more reasonable to consider the ferrite formation as a direct decomposition of austenite involving a diffusion of carbon to the graphite flakes.

41. In considering the structure of these specimens it must be borne in mind that they represent the transformation in iron heated to 1600°F., *i.e.*, austenite containing only slightly more than the eutectoid percentage of carbon. The question arises as to

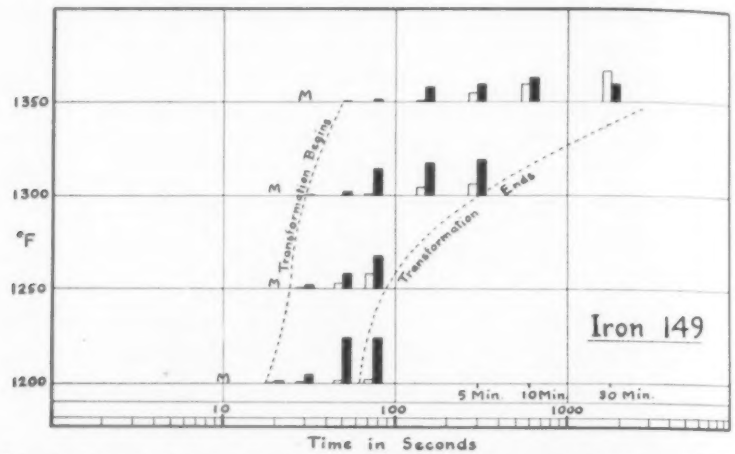

FIG. 23—SUMMARY CHART OF IRON NO. 149. BLACK BARS REPRESENT PEARLITE AND WHITE BARS FERRITE. SPECIMENS QUENCHED FROM 1800°F.

whether austenite of higher carbon content would still form free ferrite during transformation under the same circumstances, namely, a sudden cooling to subcritical temperatures. To check this, the entire series was repeated on the same irons, using an initial temperature of 1800°F.

Second Series Quenched from 1800°F.

42. Experimental procedure was the same as before, except that the specimens were heated to 1800°F. prior to quenching in

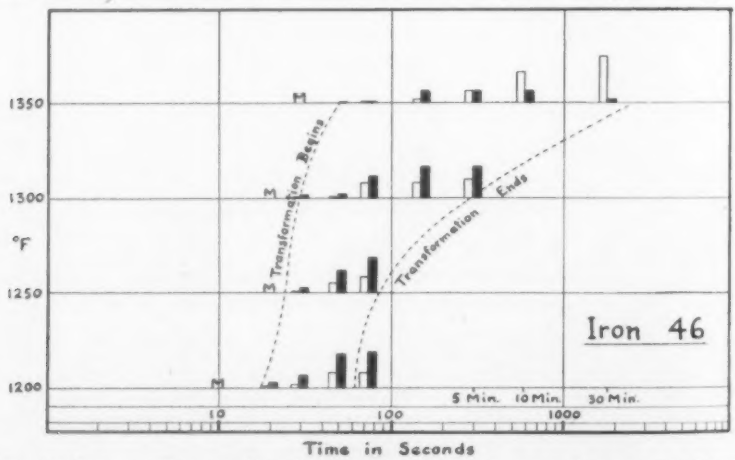

FIG. 24—SUMMARY CHART OF IRON NO. 46. BLACK BARS REPRESENT PEARLITE AND WHITE BARS FERRITE. SPECIMENS QUENCHED FROM 1800°F.

the lead pot. Eight-gram specimens were used for the whole series. String galvanometer records showed that 12 sec. were required to cool from 1800 to 1200°F. The structures observed were similar to those found in the previous series and need not be described in detail. It is sufficient to present the summary charts (Figs. 23 and 24) which were plotted in the same way as Figs. 17 and 18. These show that, while free ferrite still forms in specimens quenched from 1800°F., there is a marked difference in the relative amounts of ferrite and pearlite formed at different temperature levels.

43. In comparing the two series treated from different initial temperatures, the following points may be noted:

(1) At 1350F° and 1300°F., the rate of ferrite formation exceeds the rate of pearlite formation in specimens quenched from 1600°F. (Figs. 17 and 18). In specimens quenched from 1800°F. the reverse is the case, pearlite being formed more rapidly than ferrite (Figs. 23 and 24).

(2) At 1250°F. and 1200°F., the rate of pearlite formation exceeds the rate of ferrite formation in specimens quenched from 1600°F. This is also true for specimens quenched from 1800°F.

(3) At 1350°F. and 1300°F., transformation begins and ends in about the same time whether the initial temperature was 1600°F. or 1800°F.

(4) At 1250°F. and 1200°F., the beginning of transformation is delayed by quenching from 1800°F.

DISCUSSION OF THE STRUCTURES

44. In attempting to interpret the behavior of these specimens, some consideration may be given to the effect of undercooling on reaction rates. The fact that the metastable transformation of austenite into pearlite is accelerated at subcritical temperatures is well known. At successively lower temperatures below the equilibrium transformation range, the reaction begins and ends more quickly. The terms "tendency" or "driving force", and ease of transformation of "facility" have been used by Davenport and Bain in describing the decomposition of austenite in steel[2].

45. Taking the case of a specimen of cast iron heated to 1600°F. and suddenly cooled to 1250°F. and held at that tempera-

ture, we may attempt to follow through the mechanism of transformation using the iron, carbon, silicon equilibrium diagrams shown in Fig. 1 and Fig. 10. The matrix, at 1600°F., consists of austenite holding in solid solution an amount of carbon slightly above eutectoid composition. It is now cooled in a period of 10 sec. to 1250° F. The matrix, while still austenitic, is under-cooled far below the range in which austenite exists in equilibrium with alpha iron and graphite (the region $\alpha + \gamma + Gr$ in Fig. 10). Therefore, the "tendency" or "driving force" toward the formation of ferrite is present. A diffusion of carbon to the graphite flakes starts immediately and proceeds with great rapidity. Under such circumstances, it is conceivable that the austenite in the immediate vicinity of the graphite flakes might fall below eutectoid composition prior to the beginning of transformation. Once transformation started, ferrite would be rejected as shown in Fig. 21.

46. Meanwhile, the metastable transformation of austenite into pearlite has started at other points in the matrix. This proceeds rapidly because the specimen is undercooled far below the equilibrium transformation range for pearlite. (The region $\alpha + \gamma + + Ca$ in Fig. 1). Therefore, a competition exists between two reactions, one a direct transformation of austenite into pearlite and the other a continued diffusion of carbon to the graphite flakes with attendant development of free ferrite. It will be noted in Fig. 22-Right that the ferrite does not form continuous layers around the graphite flakes at this stage of transformation. There is still opportunity for the diffusion of carbon through austenite until the ferrite consolidates into continuous masses such as those shown in Fig. 22-Left.

47. In the case of a specimen cooled suddenly from 1600° F. to 1350° F., undercooling is not so severe. Transformation starts in the same way with the simultaneous formation of ferrite and pearlite (Figs. 17 and 18). The pearlite formation, however, proceeds at a very leisurely rate. The ferrite formation, favored by more time for diffusion, runs ahead of the pearlite, leaving the final structure predominately ferritic. The obvious breakdown of some of the pearlite in specimens held for 30 min. at 1350° F., is a secondary reaction and plays no part in the actual mechanism of transformation.

48. Specimens quenched from 1600° F. to 1300° F. in a holding bath, are intermediate in behavior to those treated at 1350° and

1250° F. In the normal iron, the ferrite and pearlite form at approximately equal rates (Fig. 17) while in the modified iron the ferrite runs ahead (Fig. 18). This difference may be attributed to the difference in graphite flake size, the modified iron having a much greater area of graphite - austenite interface, affording greater opportunity for diffusion. The gradual growth of the ferrite layers during transformation may be seen very clearly in Figs. 19 and 20.

49. In specimens quenched from 1600° F. to 1200° F. in a holding bath, the pearlite formation is greatly accelerated and runs ahead of the ferrite formation in both irons. Here again, the modified iron shows more ferrite than the normal iron at all stages of transformation (Figs. 17 and 18).

Quenching from 1800° F.

50. The case of specimens cooled suddenly from 1800° F. is somewhat different. At 1800° F., the matrix contains about 1.10 per cent carbon in solid solution, an amount well above eutectoid composition. Upon cooling rapidly to 1350° F., the proeutectoid carbon is obliged to diffuse to the graphite flakes before transformation starts. That such was actually the case is evidenced by the fact that no trace of proeutectoid carbide or free graphite could be found in the pearlite itself. Transformation started after 50 sec. with the formation of small areas of pearlite along the surface of the graphite flakes, i.e., in those places where the matrix probably reached eutectoid composition first (Figs. 23 and 24). It is evident from a consideration of Fig. 22-Right that once pearlite begins to form along the graphite-austenite interface, the subsequent development of ferrite is greatly retarded. Rapid diffusion of carbon to the graphite flakes at these points is no longer possible. Consequently, the growth of ferrite during transformation is handicapped and the greater part of the matrix transforms into pearlite. Specimens held at 1350° F. for 10 min. and 30 min. show evidence of the graphitization of pearlite, the lamellae being partly spheroidized. This again is a secondary reaction having nothing to do with the transformation proper.

51. Specimens quenched from 1800° F. to holding baths at 1300° F., 1250° F., and 1200° F. show the same general trend as those treated at 1350° F., pearlite forming more rapidly than ferrite in all cases. The delay in the beginning of transformation at

the lower temperatures may be attributed to the fact that extra time was required for the austenite to lose proeutectoid carbon by diffusion to the graphite flakes. None of these specimens shows any trace of proeutectoid carbide or free graphite in the pearlite areas.

52. The entire series of specimens quenched from 1800° F. also show a greater amount of ferrite formed in the modified iron, which may be attributed as before to the greater area of graphite-austenite interface in the iron containing fine graphite.

The Effect of Segregation

53. No mention has been made of the effect of segregation on the transformation. All the specimens examined show evidence of a relationship between the cell structure and the mechanism of transformation. In general, ferrite first appears at the center of the cells while pearlite is concentrated at the cell boundaries. This is illustrated in Fig. 25, taken at low magnification. The segregation of impurities, notably phosphorus, into the cell boundaries during the freezing of the eutectic, offers a logical explanation of this phenomenon.

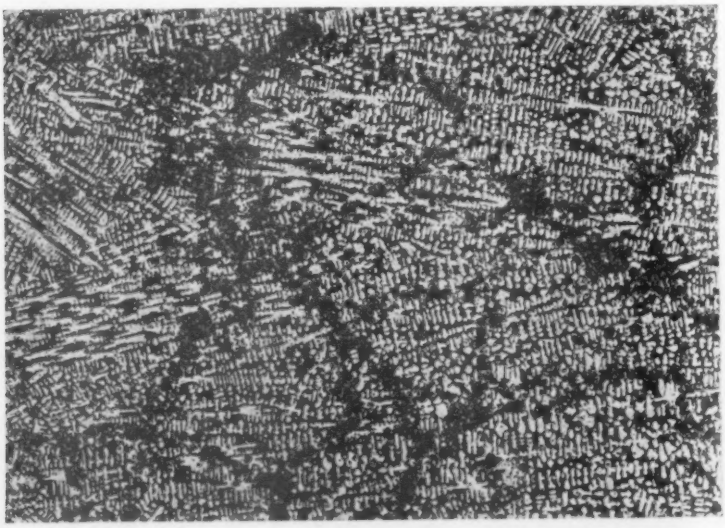

FIG. 25—TEST NO. 402—ETCHED, X20. MODIFIED IRON NO. 46 HEATED TO 1800°F. AND QUENCHED INTO A LEAD BATH STANDING AT 1200°F. HELD 30 SEC. AT 1200°F. AND WATER QUENCHED. A NETWORK OF PEARLITE HAS FORMED AT THE CELL BOUNDARIES. THE REMAINDER OF THE MATRIX IS MARTENSITE WITH SOME FREE FERRITE AT THE CENTERS OF THE CELLS.

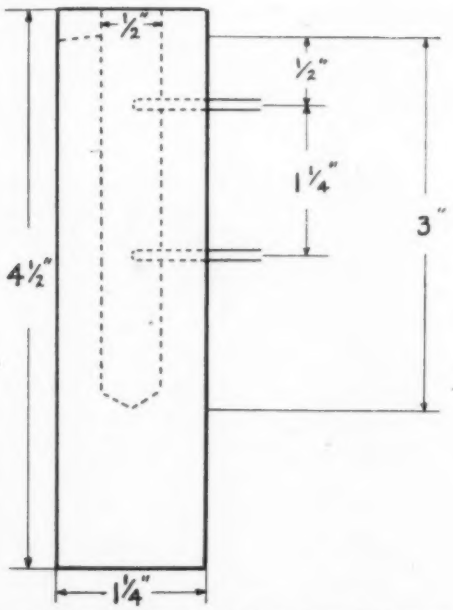

FIG. 26—DRAWING OF EXPERIMENTAL MOLD MADE FROM INSULATING BRICK, SHOWING LOCATION OF THERMOCOUPLES.

THE TRANSFORMATION IN CASTINGS

54. The thermal history of the specimens just described is quite different from that of a casting cooling continuously from the liquid state. For this reason, the progress of transformation in a series of small castings was examined by quenching at various stages of cooling in the mold.

55. After some preliminary work, the type of mold shown in Fig. 26. was adopted. These were made from insulating brick, by drilling a hole of ½-in. diameter by 3¼-in. deep. The outside was cut accurately cylindrical to give a uniform wall thickness of ⅜-in. Thermocouples were cemented in place, as shown, with the hot junction at the center of the bar. Chromel-alumel, 22-gauge wire was used, with the bead protected by a thin coating of porcelain cement. A runout, provided at the top of the mold, limited the amount of metal above the top thermocouple to ½-in. The extremely porous nature of the insulating brick made it possible to quench the casting at any instant during cooling by plunging the entire mold into water. Specimens quenched from above the transformation range were fully martensitic with no trace of

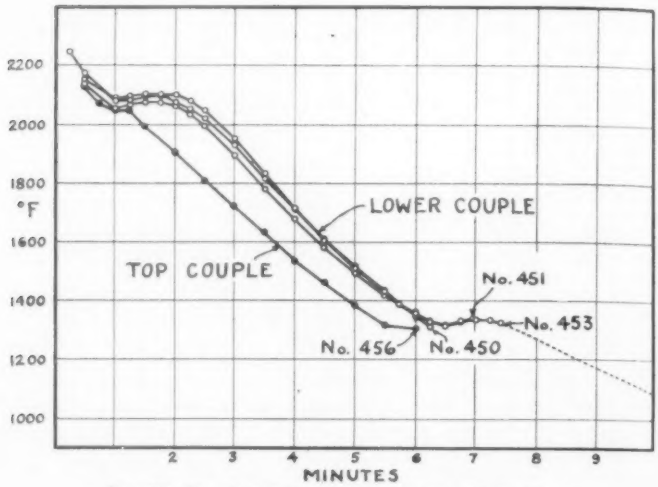

FIG. 27—COOLING CURVES OF EXPERIMENTAL CASTINGS.

troostite, indicating that the method was suitable for arresting the progress of transformation at any stage with the assurance that any remaining austenite would appear as martensite in the final structure.

56. Normal gray iron of the following analysis was used as melting stock: total carbon, 2.96 per cent; silicon, 2.12 per cent; manganese, 0.68 per cent; sulphur, 0.064 per cent; phosphorus, 0.098 per cent.

57. Charges, machined from transverse test bars, were melted in an alundum crucible supported in a graphite holder in a globar-tube furnace. It is known from a great deal of previous work, that this melting precedure gives practically no change in analysis of the melting stock and has the further advantage of giving accurately known temperatures. The melting and pouring temperature was 2500°F. in all cases. Starting from the time of pouring, temperature readings were taken down to the instant of quenching. Several of the curves are given in Fig. 27.

Structure of the Castings

58. Each casting was split vertically so that the entire piece could be examined from top to bottom. From the structures, it was evident that most of the heat flow during cooling was toward the top of the casting. Traces of chill occurred at the very top surface. Below this was very fine graphite with considerable fer-

rite (Fig. 29). The structure was practically normal at a depth of ½-in. below the top (Fig. 30-Left) and the graphite flakes gradually increased in size at greater depths (Fig. 30-Right). The structure, as regards graphite size at any given distance from the

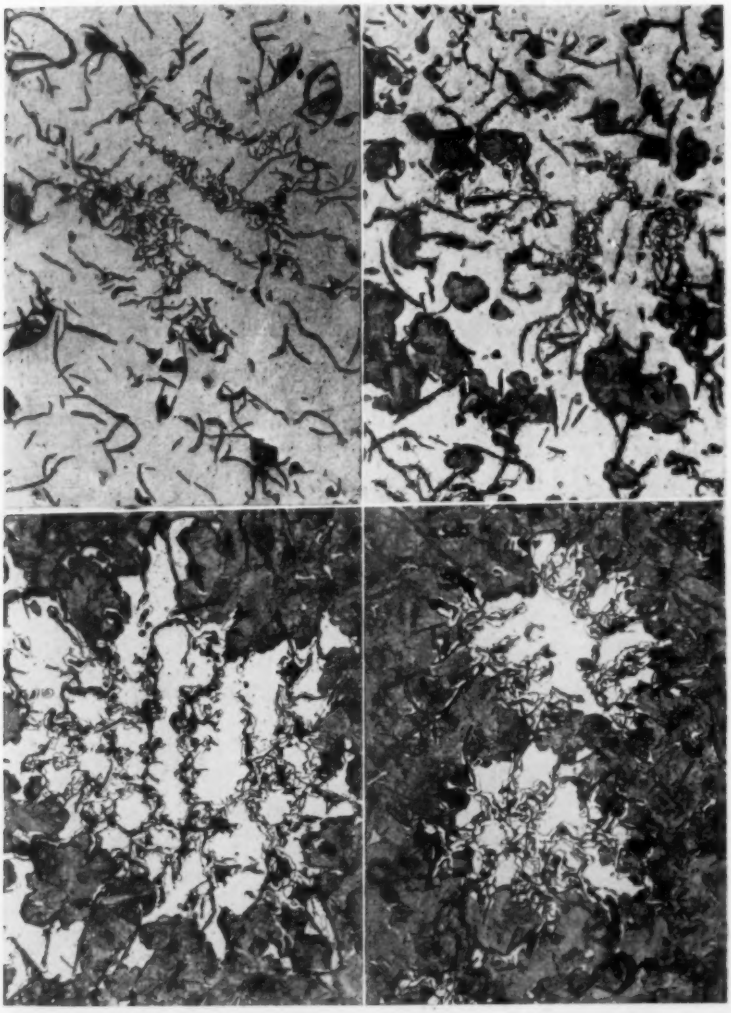

FIG. 28—TOP LEFT—CASTING NO. 456. 1¾-IN. FROM TOP. ALL MARTENSITE. ETCHED, X100. TOP RIGHT—CASTING NO. 450. 1¾-IN. FROM TOP. FERRITE + PEARLITE + MARTENSITE. ETCHED, X100. BOTTOM LEFT—CASTING NO. 451. 1¾-IN. FROM TOP. FERRITE + PEARLITE + MARTENSITE. ETCHED, X100. BOTTOM RIGHT—CASTING NO. 453. 1¾-IN. FROM TOP. FERRITE + PEARLITE + MARTENSITE. ETCHED, X100.

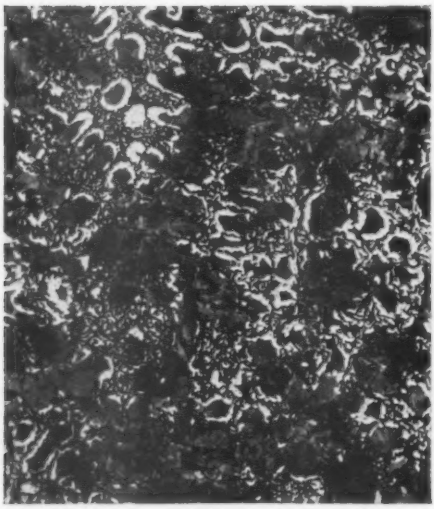

FIG. 29—CASTING NO. 451. ¼-IN. FROM TOP. ETCHED, x100.

top, was quite uniform in all the castings. Therefore, attention will be confined to the progress of transformation during the pearlite interval, assuming that the specimens represent a uniform material arrested at various stages of cooling in the mold.

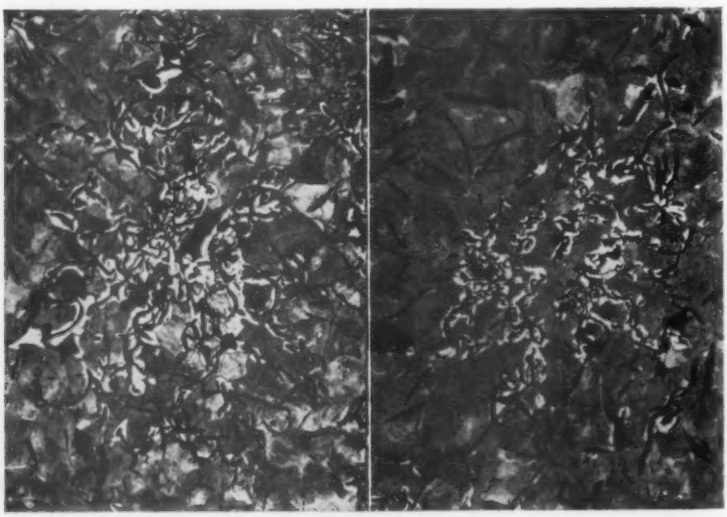

FIG. 30—LEFT—CASTING NO. 457. ½-IN. FROM TOP. ETCHED, x100. RIGHT—CASTING NO. 457. 1¾-IN. FROM TOP. ETCHED, x100.

FIG. 31—CASTING NO. 456. 1½-IN. FROM TOP. EARLY STAGE OF TRANSFORMATION SHOWING FERRITE NEEDLES ADJACENT TO THE GRAPHITE FLAKES, ETCHED, X500.

Test No. 456

59. This casting was quenched just after the top couple began to show an arrest (1310°F.). The structure adjacent to the lower couple (Fig. 28-Top Left) consists of martensite and graph-

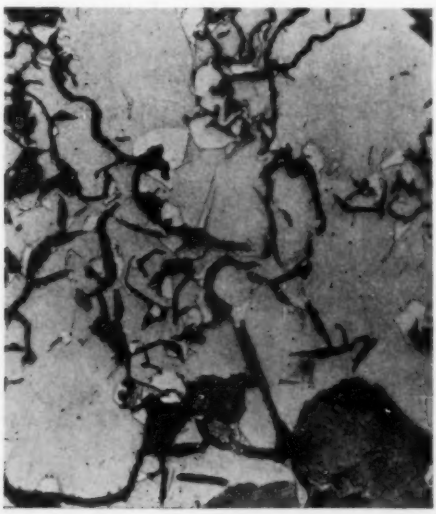

FIG. 32—CASTING NO. 456. ½-IN. FROM TOP. LATER STAGE OF FERRITE DEVELOPMENT. ETCHED, X500.

ite. The existence of a fully martensitic matrix is evidence that the quenching rate was sufficiently rapid to prevent the formation of any lamellar constituents which might be confused with the pearlite in specimens quenched at later stages of cooling. Fig. 32 shows the structure adjacent to the upper couple. Here transformation is under way as indicated by the cooling curve (the lower curve of Fig. 27). Ferrite has already formed along the small graphite flakes at the cell centers. Patches of pearlite occur at the cell boundaries. Between the two locations, shown in Figs. 28-Top Left and 32, all intermediate stages of transformation are present. Fig. 31, taken ¼-in. above the lower couple, shows an early stage in the ferrite formation.

60. Ferrite first appears as fine needles oriented along certain planes in the austenite immediately adjacent to the graphite flakes. These grow in width and become consolidated into masses of ferrite, many of which show sharp angles and points (Fig. 32). There is no indication that such ferrite results from a breakdown of previously existing pearlite. Its rejection along certain crystallographic planes in the austenite resembles the rejection of ferrite in steels, forming the well known Widmanstätten structure. Apparently the ferrite formation begins at a very early stage in the transformation, at a time when no trace of pearlite is present.

Test No. 450

61. This specimen was quenched about 15 sec. later than No. 456 (See Fig. 27). Fig. 28-Top Right shows the structure adjacent to the lower couple. A considerable amount of ferrite occurs at the cell centers. Pearlite is forming as small patches scattered through the matrix.

Test No. 451

62. This specimen was quenched 45 sec. after the beginning of the eutectoid arrest on the lower couple (Fig. 27). Fig 28-Bottom Left shows the structure adjacent to the lower thermocouple junction. Transformation is well advanced. What austenite remains is located mainly in the cell centers. The ferrite in these locations still shows needlelike and angular forms.

Test No. 453

63. This specimen was quenched near the end of the eutectoid arrest (Fig. 27). The structure adjacent to the lower couple

is shown in Fig. 28—Bottom Right. Small patches of austenite remain at the cell centers. The ferrite in these locations has lost its needlelike appearance and is apparently fully developed. At a short distance above the position shown, transformation was complete.

Test No. 457

64. This casting was slow cooled in the mold to room temperature. The structure adjacent to the lower couple is shown in Fig. 30—Right. It consists of pearlite with free ferrite in the cell centers. The structure ½-in. from the top is shown in Fig. 30—Left. At ¼-in. from the top, the structure was identical to Fig. 29

65. Four additional castings were made duplicating those listed above. These showed the same stages of transformation and need not be described.

PROGRESS OF TRANSFORMATION IN CASTINGS

66. The progress of transformation in these castings resembles that observed in the specimens quenched in lead. Transformation occurred at about 1330°F. and required something over a minute to reach completion (Fig. 27). This rate of transformation, under conditions of continuous cooling, is about equal to that of the specimens quenched into lead standing at 1250°F. and held at constant temperature. (See Figs. 17, 18, 23, and 24). Using the general ideas expressed concerning the specimens quenched in lead, we may attempt to form a mental picture of what happened during the cooling of these castings, starting with the solidified metal at a temperature of 2000°F.

67. At 2000°F., the matrix consisted of austenite holding in solid solution about 1.50 per cent carbon. As the metal cooled in the mold, a part of the carbon in solution diffused to the graphite flakes and the austenite approached eutectoid composition. In order for diffusion to occur, a composition gradient must have existed, for example, the austenite at point *"A"* in Fig. 31 must at all times have been higher in carbon than the austenite at point *"B"*. If the cooling had been slow enough, the matrix, according to Fig. 1, would have started to transform into pearlite at about 1470°F. Actually, the degree of undercooling was such that transformation started at 1325°F. This temperature is well

below the range in which ferrite, austenite and graphite exist in equilibrium (Fig. 10). Therefore, the "tendency" or "driving force" toward the formation of ferrite is present.

68. Under such circumstances, there is no reason why the austenite at point "*B*" in Fig. 31 should not continue to diffuse carbon to the graphite flake nearby, falling momentarily below eutectoid composition. As a result, free ferrite would appear in the neighborhood of the graphite flakes, the phases alpha, gamma, and graphite existing side by side. The first appearance of ferrite, as needles arranged along the crystallographic planes of the austenite, is strongly suggestive of a hypoeutectoid reaction.

69. It is known from the specimens quenched in lead that the rate of formation of ferrite is accelerated at subcritical temperatures. Ferrite appeared after 30 sec. in specimens cooled suddenly from 1800 to 1300°F. (Fig. 24). The matrix of Fig. 31, which is undercooled to 1325°F., has had ample time to form the amount of ferrite shown. The specimens quenched in lead also show that ferrite continues to form during transformation (Figs. 19 and 20). At the same time, the metastable transformation of austenite into pearlite is going on at other points in the matrix. Therefore, a competition exists between the two reactions, both of which are accelerated by undercooling. Once the ferrite needles in Fig. 31 grow into continuous masses, any further diffusion of carbon to the graphite flakes must occur through alpha iron. As transformation progresses, the ferrite formation slows down while the pearlite formation speeds up, heat being drained continuously from the casting.

70. There is no evidence in these specimens that any ferrite forms after transformation is complete. Castings cooled in the mold to room temperature do not show any more free ferrite than those quenched in the last stages of transformation.

71. Whether free ferrite or pearlite begins to form first at any given point in the matrix is not entirely a matter of chance. Ferrite is distinctly concentrated at the center of the cells formed during the freezing of the eutectic, which are the places containing the smallest graphite flakes or, otherwise expressed, where the area of the graphite-austenite interface is a maximum. The pearlite formation is concentrated at the cell boundaries exactly as was found in the specimens quenched in lead (Fig. 25). Aside from

the difference in flake size, the segregation of impurities into the cell boundaries, during freezing, probably influences the subsequent transformaton in these places.

72. The influence of the size of the graphite flakes may be seen by comparing Fig. 29 with Fig. 30—Left. The matrix in Fig. 30—Left transformed at 1300°F. Although the matrix in Fig. 29 must have transformed at a lower temperature, it was nevertheless subject to the same influences, namely, an acceleration of both the stable and metastable reaction by undercooling. Not only is the total area of graphite-austenite interface greater in Fig. 29, but the mean spacing of the graphite flakes is much less. At the same time, the number of cell centers, which are favorable points for ferrite formation, is greater because of the faster rate of freezing. All these factors should contribute to a greater total quantity of free ferrite. The structures, shown in Figs. 29 and 30—Left, are analogous to those of the normal and modified irons quenched into lead, which showed, for any given treatment, a greater amount of free ferrite in the iron containing fine graphite.

SUMMARY

73. The formation of free ferrite during transformation in gray iron containing from 2.93 per cent to 3.03 per cent total carbon and 2.12 per cent to 2.34 per cent silicon was investigated as follows:

1. Specimens annealed in vacuo for 96 hr. and water quenched, indicated that ferrite, austenite and graphite exist in equilibrium over a range of temperature extending from 1450°F. to 1550°F.

2. Specimens were heated to 1600°F. and quenched into a lead bath standing at constant subcritical temperature, held for various lengths of time and finally quenched in water. These showed that the rate of formation of both pearlite and ferrite was accelerated at subcritical temperatures over the range covered (1350 to 1200°F.). Under identical conditions of heat treatment, an iron containing fine graphite flakes showed more ferrite than one containing large graphite flakes. This was attributed to the difference in total area of graphite-austenite interface.

3. The same work was repeated using an initial temperature of 1800°F. The rate of formation of free ferrite was still accelerated at subcritical temperatures, but less ferrite formed in any given treatment. Iron containing fine graphite flakes still showed more ferrite than iron containing large graphite flakes.

4. The progress of transformation in a series of small castings was examined by quenching at various stages of cooling in the mold. It was found that ferrite started to form along the graphite flakes prior to the formation of pearlite and continued to develop during the transformation period. No additional ferrite appeared after transformation was complete.

74. It was concluded, from these experiments, that silicon not only promotes graphitization but also provides a mechanism for the formation of free ferrite by causing the phases, alpha, gamma, and graphite to exist in equilibrium over a range of temperature well above the pearlite interval.

ACKNOWLEDGMENTS

75. The work described above was carried out as a part of the program of fundamental research at Battelle Memorial Institute.

76. The writer wishes to express his appreciation for the advice and counsel received from various members of the staff, particularly to Dr. C. H. Lorig under whom this work was conducted. He also desires to express thanks to Arthur R. Elsea for his capable assistance in the experimental work.

Appendix

Method of Estimating the Amounts of Ferrite and Pearlite

Several prints, such as Fig. 5—Left, were cut to circular form and weighed on an analytical balance. With small scissors the primary dendrites were then cut out and weighed. Five determinations ranged from 55 to 65 per cent for the area of the dendrites. In the modified iron, it was concluded that the filling material represented 40 per cent of the average area of a field. Of this amount, fully half is graphite (See Figs. 21 and 22—Left). Therefore, the value 20 per cent was taken as representing the average area of graphite in a field.

To check this, the graphite flakes were cut from a number of prints of normal iron. These gave values ranging from 18 to 21 per cent. In the normal iron, the amount of ferrite could be estimated by comparing it to the graphite flakes. In the modified iron, it was judged from the amount in the dendritic fillings (a maximum of 20 per cent) plus the degree of encroachment on the dendrites.

The percentages plotted in the summary charts are estimates made by sweeping over the whole specimen not counting the decarburized skin at the surface. They represent the per cent by area in the average field, no attempt being made to convert to volume or weight. The phosphide eutectic was ignored in making these estimates.

It was found advantageous to use polarized light in looking for the first traces of pearlite formed during transformation. By crossing the polarizer and analyzer, the tiny patches of pearlite stand out brilliantly on a dark background. In specimens to be photographed, the martensite was tempered very slightly (one minute at 500°F.) to make it etch more rapidly and provide a better contrast with the free ferrite.

Literature on the Austenite Transformation in Gray Iron

The diagrams shown in Fig. 1, were drawn from all existing information on the iron, carbon, silicon system, references to which may be found in the original monograph.[1] Murphy and Wood, and co-workers have published several papers on austenite transformation in gray iron [3,4,5,6]. These authors were concerned mainly with hardenability as interpreted in the light of the "*S*" curve. The practical aspects of heat treatment have recently been described in an excellent paper by Timmons, Crosby and Herzig[7] who give references to earlier work on the subject. These authors used the iron, carbon, silicon diagram in discussing their results and demonstrated the rapidity of graphitization in irons heated to temperatures falling in the region $\alpha + \gamma + Gr$. Earlier references to the formation of ferrite in castings are those of Bancroft and Dierker[8] and of Parke, Crosby and Herzig[9].

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7. Timmons, G. A., Crosby, V. A., and Herzig, A. J., *"Some Factors Involved in Hardening and Tempering Gray Cast Iron,"* TRANSACTIONS, American Foundrymen's Association, vol. 47, pp. 397-417 (1939).

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DISCUSSION

Presiding: S. C. MASSARI, Association of Manufacturers of Chilled Car Wheels, Chicago, Ill.

CHAIRMAN MASSARI: Will Mr. Boyles briefly describe the application of these basic data to our every day problems?

MR. BOYLES: Referring to Fig. 16, something might be said about heat treatment. Suppose, for example, that you wanted to harden a piece of cast iron of 3 per cent carbon and 2 per cent silicon. It is evident that the iron would have to be heated above the upper line of the region $\alpha + \gamma + Gr$ to get the austenite above eutectoid composition. As soon as the metal crossed the lower line at 1450°F., ferrite would form very rapidly and heating would have to be continued to some point above the line at 1550°F. to produce a fully austenitic matrix. According to this diagram no amount of holding at temperatures between 1450 and 1550°F. would ever produce a fully austenitic matrix. Specimens

quenched from temperatures between these limits would consist of a mixture of ferrite and martensite, *i.e.* the material would not be fully hardened.

The paper also offers a possible explanation for the formation of the so called "primary" ferrite in castings. The work indicates that the rate of formation of ferrite during cooling is influenced by the size of the graphite flakes. With a given composition, more ferrite should form in a casting having fine graphite flakes as compared to one having coarse graphite flakes. This effect is especially noticeable in wedge tests where a region near the tip, which contains very fine graphite, usually shows more ferrite than the heavier parts of the wedge.

CHAIRMAN MASSARI: When subjecting a casting to double heat treatment to accomplish increased hardness, if the casting is heated in the range referred to by Mr. Boyles, you of course produce a substantial percentage of ferrite, which will inhibit the subsequent hardening of the casting. One might heat treat a casting by heating it to this temperature range rather than beyond it and on quenching expect to substantially harden it and in reality obtain only a rather mediocre result.

MEMBER: Is it possible by an examination of the microstructure to determine the range in which the ternary eutectic of carbide and phosphide is broken down? I do not mean the pearlite, but the massive carbides.

MR. BOYLES: I should think that could be determined by heating specimens in that temperature range.

MEMBER: Did the author notice any particular range in which these carbides begin to diminish?

MR. BOYLES: I did not look for that in this work. The phosphorus is rather low in these irons and that did not occur to any extent.

MEMBER: I would like to ask Mr. Boyles if he has an opinion on the effect of variations in silicon content upon the precise position of the transformation range?

MR. BOYLES: Fig. 1 shows three sections through the ternary diagram of the iron-carbon-silicon system. Notice that the range $\alpha + \gamma + Ca$ becomes wider as the silicon content increases. These sections lie right behind one another in the solid model of the system and the region in question tapers out like a wedge from the iron-carbon side. There is a definite relationship between the silicon content and the range of temperature in which transformation occurs. This diagram represents the metastable condition. So far as I know, there has been no determination of the range in which alpha, gamma, and graphite exist in equilibrium in alloys covering a wide range of silicon contents.

MEMBER: Referring to Paragraph 74, which states, "but also provides a mechanism for the formation of free ferrite by causing the phases, alpha, gamma, and graphite to exist in equilibrium over a range

of temperature well above the pearlite interval." Will you explain that?

MR. BOYLES: Assuming Fig. 1 is correct for the metastable transformation the range shown on this diagram is considerably lower than the temperatures which I found by long annealing and quenching. You notice the lower line comes a little above 1400°F. and the upper line comes below 1500°F. for 2 per cent silicon and 3 per cent carbon.

MEMBER: I would like to ask Mr. Boyles about a statement in paragraph 50 where he says—"Upon cooling rapidly to 1350°F., the proeutectoid carbon is obliged to diffuse to the graphite flakes before transformation starts." Do you think that transformation cannot start until we have reached the eutectoid composition?

MR. BOYLES: The statement means that the austenite cannot transform until it reaches approximately eutectoid composition.

DR. R. W. LINDSAY:[1] In paragraph 43 the following statement is made. "At 1350 and 1300°F., the rate of ferrite formation exceeds the rate of pearlite formation in specimens quenched from 1600°F. In specimens quenched from 1800°F., the reverse is the case, pearlite being formed more rapidly than ferrite." I would like to ask Mr. Boyles' opinion as to the reason for this difference in behavior. Would he attribute it to a difference in carbon solubility in the austenite at the eutectoid transformation range resulting from carbon solubility differences at 1600 and 1800°F.?

MR. BOYLES: I think that is a matter of relative rates of transformation in two systems, stable and metastable. That is, in the higher temperature ranges, the stable transformation is more rapid than the metastable, while in the lower temperature ranges, the reverse is the case.

DR. R. W. LINDSAY: When those samples were quenched from 1600°F. to 1300 and 1350°F., you found that the rate of ferrite precipitation was more rapid than the rate of pearlite formation, and when the samples were quenched from 1800°F. to the same temperatures, the rate of pearlite formation was more rapid than the rate of ferrite formation.

MR. BOYLES: I think it is due to the time required for the austenite to lose carbon by diffusion to the graphite flakes. I looked very carefully for any traces of free graphite or carbide in the matrix and there was no trace of either. Apparently all carbon goes to the flakes. A certain amount of time was required for the austenite to drop to the proper carbon concentration and that interfered with and slowed up the formation of ferrite.

MEMBER: I would like to ask Mr. Boyles in what manner he confirmed metallographically the fact that the needles he showed in Fig. 31 were ferrite? They somewhat appeared like carbide flakelets.

[1]Research Assistant, Association of Manufacturers of Chilled Car Wheels, Chicago, Ill.

MR. BOYLES: As near as I could tell, they had the metallographic characteristics of ferrite. That is, they scratched easily and did not blacken with alkaline sodium picrate. In addition, in many cases, they were joined to and tapered from grains which were large and which were obviously ferrite.

Effect of Manganese on Second-Stage Graphitization

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Rockford, Ill.

Abstract

The authors investigated the effect of manganese on the graphitizing properties of white iron by using ferromanganese ladle additions to give manganese percentages of 0.45, 0.91, and 1.28. All samples were given a uniform first-stage graphitization treatment and then heated at 1300°F. for periods from 45 min. to 48 hr.

Progress of second-stage graphitization was followed by noting reduction of combined carbon at the end of various periods of time. As manganese increases, the rate of graphitization is greatly reduced.

The authors discuss graphitization under various phase conditions and note evidence of the influence of manganese in other phases than the one covered by the title of the paper.

1. Many individuals have contributed to the knowledge of the mechanism of converting the combined carbon of white cast iron into the temper carbon of malleable iron. Moldenke, Touceda and Schwartz, among many others, have written at length on this subject.

2. Much of the early information has been revised in the light of later experience and more complete and painstaking investigation. It now appears that the malleablizing process is a very complex one because of the many variables influencing the process.

3. We mention, briefly, some of these variables which must be considered before malleablizing can be reduced from an art to a science:

(1) the "history" of the molten metal prior to casting;

(2) the cooling rate, or time-temperature cycle, during solidification and cooling to room temperature;

* President and Metallurgists, respectively, Gunite Foundries Corp.

Note: This paper was presented before the Malleable Cast Iron Session, May 7, 1940, at the 44th Annual A.F.A. Convention in Chicago, Ill.

(3) the grain size and grain distribution of the resulting white iron;

(4) the analysis of the iron, including metallic, non-metallic, and gaseous elements;

(5) the rate of reheating to the graphitization temperature;[1*]

(6) the time and temperature used for first stage graphitization;

(7) the time-temperature cycle used in effecting second stage graphitization, and

(8) oxidizing, decarburizing, or carburizing conditions in the atmosphere surrounding the castings.

4. A malleable iron foundryman is faced with additional variables brought about by conditions of commercial operations; because of annealing furnace size, castings located in various parts of the furnace do not follow exactly the same time-temperature cycle and each furnace load contains castings which are not of identical analysis, grain size, and solidification rate.

5. Fortunately for the malleable industry, these variations become unimportant, provided an annealing cycle is selected which is long enough in the first and second graphitization stages so that graphitization goes to completion, even for metal of the poorest graphitization characteristics. As long as graphitization goes to completion, the metal with rapid graphitizing properties comes from the oven in practically the same form as the metal in which graphitization is slower.

6. Twenty years ago, from 7 to 12 days were required for safe annealing. Now, thanks to better furnaces, closer control of metal uniformity, and more complete knowledge of the process, white iron of standard analysis can be graphitized in something under 36 hr., and irons of "special" analysis can be annealed in 15 hr. or less[2].

7. "Pearlitic malleable" differs from malleable iron in that a portion of the carbon (usually 0.40 to 0.80 per cent) is retained as combined carbon in the finished product. This is accomplished in two principal ways: (1) by reheating completely malleablized

* Superior numbers refer to bibliography at end of paper.

iron to redissolve a portion of the temper carbon, followed by a heat treatment to develop the desired structure; or (2) by stopping the second stage graphitization of carbon before it has gone to completion, using a heat treating cycle which will produce the desired final structure.

8. The purpose of this paper is to consider features of the latter process, particularly with reference to the effect of manganese as a graphitization retardant and carbide stabilizer.

GRAPHITE PRECIPITATION

9. Graphite in iron may form under at least four phase conditions: (1) from liquid iron during solidification; (2) from austenite in the presence of massive cementite; (3) from supersaturated austenite after massive cementite has been dissolved and (4) from cementite in alpha iron, namely pearlite.

10. The various alloys which influence graphitization in gray and white iron appear to have different relative graphitizing powers under each of these four conditions. Therefore, to obtain a complete understanding, the influence of each alloying element must be studied in relation to each of these conditions.

11. The first condition (graphitization from the liquid) need not concern us because we are considering in this paper only iron which solidifies as white iron.

12. The second condition was discussed in a previous paper[3], and it was concluded that variation in manganese content, within reasonable limits, had little, if any, effect on graphitization rates during first stage graphitization. Others[4] have reached this same conclusion. However, evidence uncovered in this investigation indicates that first stage graphitization is retarded to an appreciable extent by high manganese contents.

13. The third condition is encountered at the end of first stage graphitization, namely, after massive cementite has been dissolved and the excess carbon, not soluble in austenite, has been precipitated as temper carbon. It is then necessary to lower the temperature to a point where the excess carbon in the austenite will also precipitate as temper carbon.

14. As soon as all the cementite has been dissolved in the austenite, the metastable iron-carbon diagram no longer governs;

and equilibrium conditions depend on the stable iron-carbon relationship. If we assume an iron at 1700°F. in which this condition exists and cool this iron so that the percentage of combined carbon in solution in the austenite is greater than the A_{gr} line indicates is soluble in the austenite at the lower temperature, then additional temper carbon will be precipitated to restore equilibrium.

15. Time is required to restore this equilibrium because the carbon in the austenite must migrate to the temper-carbon nodules. A carbon gradient is set up, with stable equilibrium concentration at the austenite-temper carbon interfaces, and super-saturation at extreme distances from the temper carbon nodules.

16. If the temperature is lowered too rapidly, the carbon content of the austenite may exceed the saturation point of austenite under metastable equilibrium conditions. In other words, the A_{cm} line may be crossed for the portions of the metal having greatest carbon concentration. If this occurred, one would expect precipitation of cementite similar to that found in hyper-eutectoid steel.

17. Evidence was found in this investigation that manganese influences the safe rate of cooling, which will prevent cementite precipitation of this nature. The bars of higher manganese content used in this investigation probably contained some of this cementite because the ductility of the bars, after second stage graphitization, were somewhat lower than previous commercial production led the authors to expect.

18. Concerning the fourth condition of graphite precipitation, namely, from cementite in alpha iron, the previous paper observed "that the function of manganese as a graphitization retardant was one of slowing up the graphitization of combined carbon in pearlite . . . below the critical."

19. The present investigation was undertaken principally to study the effect of varying manganese percentages in the elimination rate of combined carbon in an iron heat-treated for varying periods of time at a temperature below the critical. The temperature chosen was 1300°F., partly because it was quite close to the critical and partly because pearlite spheroidizes at this temperature, and the information gained, would have useful commercial value.

TEST PROCEDURE

20. Standard malleable-type tensile bars were cast from air-furnace, white iron of the following base analysis:

Table 1

ANALYSIS OF STANDARD MALLEABLE TENSILE TEST BARS

	Per Cent
Carbon	2.30
Silicon	0.98
Manganese	0.35
Phosphorus	0.141
Sulphur	0.092

21. By means of ferromanganese ladle additions, three sets of bars were cast containing different percentages of manganese. All bars of each set were cast from the same ladle. The analysis of each set of bars is given in Table 2, as determined in the white iron state.

Table 2

CHEMICAL ANALYSIS OF TEST BARS

Set No.	*Mn.*	*Si.*	*T.C.*	*S.*	*P.*
	%	%	%	%	%
1	0.45	0.98	2.30	0.09	0.14
2	0.91	1.00	2.28	0.09	0.14
3	1.28	0.98	2.28	0.09	0.14

22. All of the bars were heat treated for first stage graphitization by heating to 1720°F. in 18 hr., holding at 1720°F. for 30 hr., and cooling to room temperature in still air in about 1 hr.

23. Three bars from each set were assembled into four groups.

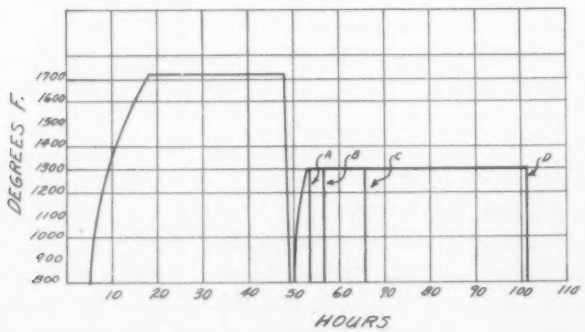

FIG. 1—CHART SHOWING TIME-TEMPERATURE CYCLE USED IN HEAT TREATING TEST BARS.

making nine bars to a group, or 36 bars in all. All the bars were then reheated to 1300°F. and held at this temperature. At the end of 45 min., the group *A* bars were removed from the furnace and cooled in air. The group *B* bars were removed at 3 hr.; the group *C* bars at 12 hr.; and the group *D* bars at 48 hr. The time intervals were selected so that each succeeding group of bars was held at temperature four times as long as the preceding group. The time-temperature cycle is shown in Fig. 1.

RESULTS

24. Combined carbon content of the bars was determined after heat treatment. Tensile properties were determined in accordance with standard procedure for malleable iron. The results of like bars were averaged and the averages shown in Table 3.

DISCUSSION OF RESULTS

25. As will be noted from Table 3, the combined carbon content of the specimens containing 0.91 and 1.28 per cent manganese, after heat treatment for 45 min. at 1300°F., was considerably in excess of that shown for the specimens containing 0.45 per cent manganese Fig. 2.

26. Examination of microscopic specimens cut from group *C* bars (1.28 per cent manganese) showed scattered traces of unde-

Table 3

RESULTS OF TESTS SHOWING AVERAGES TAKEN FROM TEST BARS AFTER HEAT TREATMENT AT 1300°F.

Hours at Temp.	*Mn. %*	*Combined Carbon, %*	*Ult. Strength, lb. per sq. in.*	*Yield Strength, lb. per sq. in.*	*Elong. in 2-in., %*
¾	0.45	0.69	89,250	57,600	4.5
	0.91	0.94	103,400	58,300	3.5
	1.28	0.96	103,350	58,500	3.25
3	0.45	0.68	83,750	56,800	4.0
	0.91	0.91	106,500	60,550	6.5
	1.28	0.98	102,500	64,100	4.5
12	0.45	0.42	80,400	46,300	8.75
	0.91	0.72	96,150	57,900	7.0
	1.28	0.92	103,000	65,100	7.25
48	0.45	0.065	56,700	36,660	20.0
	0.91	0.44	80,700	52,000	10.0
	1.28	0.66	93,500	58,560	7.0

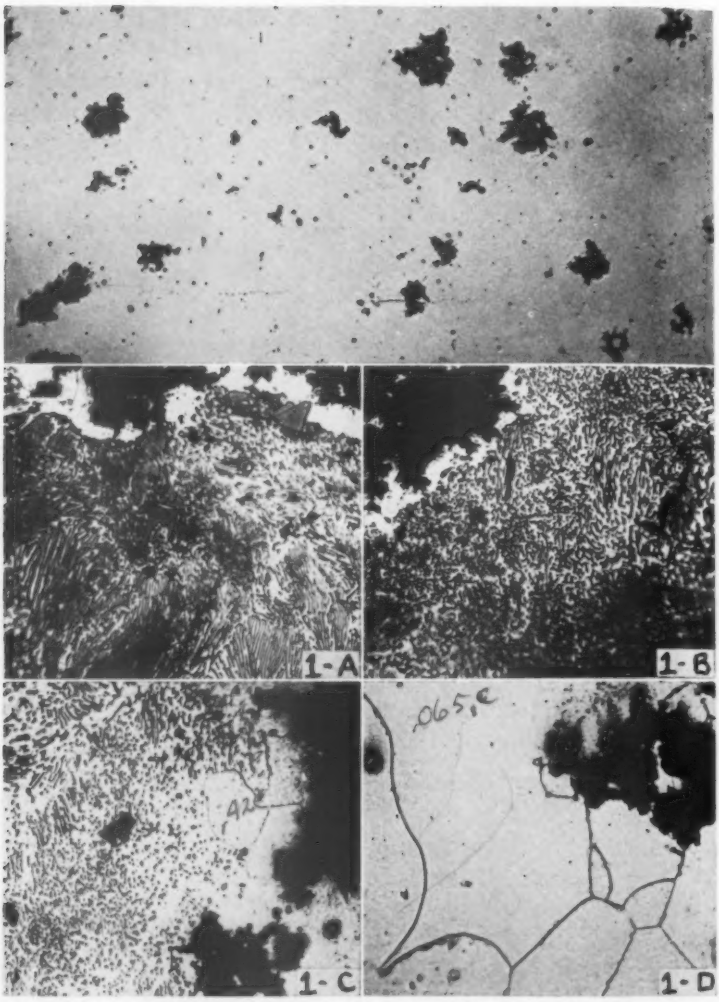

FIG. 2—SET 1. PHOTOMICROGRAPHS OF A 0.45 PER CENT MANGANESE TEST BAR HEAT TREATED AT 1300°F. FOR VARIOUS LENGTHS OF TIME. (TOP) HEATED FOR 45 MIN. UNETCHED. X100. (CENTER LEFT) 1-A HEATED FOR 45 MIN.; 1-B HEATED FOR 3 HR.; 1-C HEATED FOR 12 HR.; AND 1-D HEATED FOR 48 HR. ETCHED IN 3 PER CENT NITAL. X500.

composed massive cementite. It is, therefore, believed that manganese over 1.00 per cent does definitely retard the solution of cementite in austenite, either by stabilizing the cementite itself or by slowing up the carbon migration rate so that the austenite can not carry the carbon away so fast.

27. It is also believed that high manganese content of the

bars in Figs. 3 and 4 (sets 2 and 3) caused a lag in the expected reduction of combined carbon to the eutectoid concentration during the rather rapid cooling from 1720°F. to room temperature. It was also thought that a hyper-eutectoid matrix existed in these specimens prior to the heat treatment at 1300°F. Apparently, 45 min., or even 3 hr., at 1300°F. did not cause any great reduction in the combined carbon concentration of the higher manganese bars.

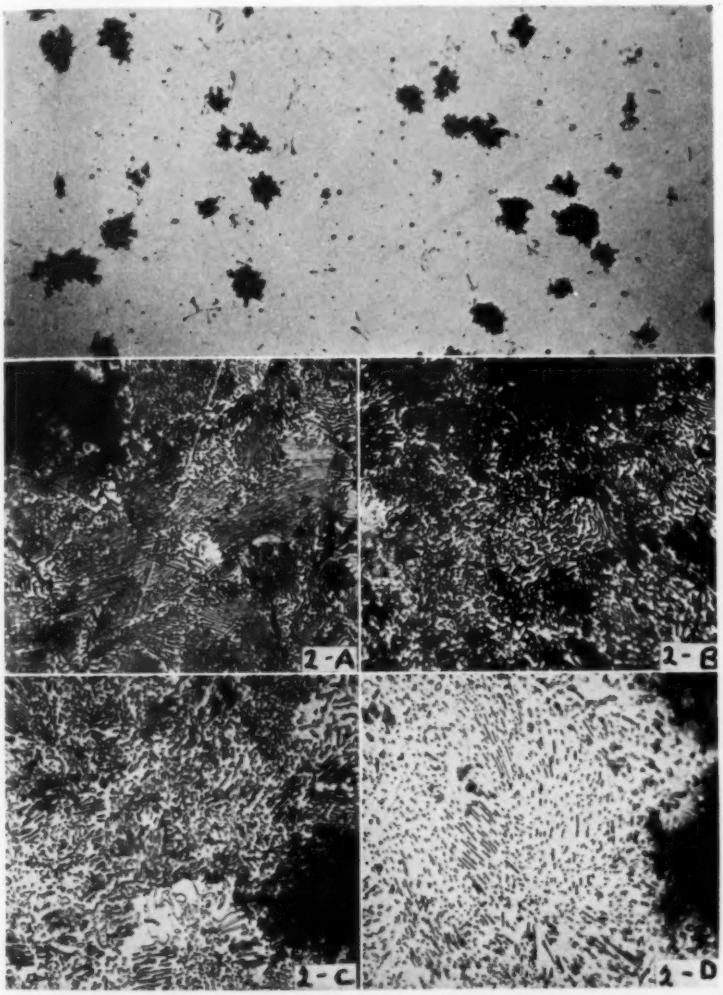

FIG. 3—SET 2. PHOTOMICROGRAPHS OF A 0.91 PER CENT MANGANESE TEST BAR HEAT TREATED AT 1300°F. FOR VARIOUS LENGTHS OF TIME. (TOP) HEATED FOR 45 MIN. UNETCHED. X100. (CENTER LEFT) 2-A HEATED FOR 45 MIN.; 2-B HEATED FOR 3 HR.; 2-C HEATED FOR 12 HR.; AND 2-D HEATED FOR 48 HR. ETCHED IN 3 PER CENT NITAL. X500.

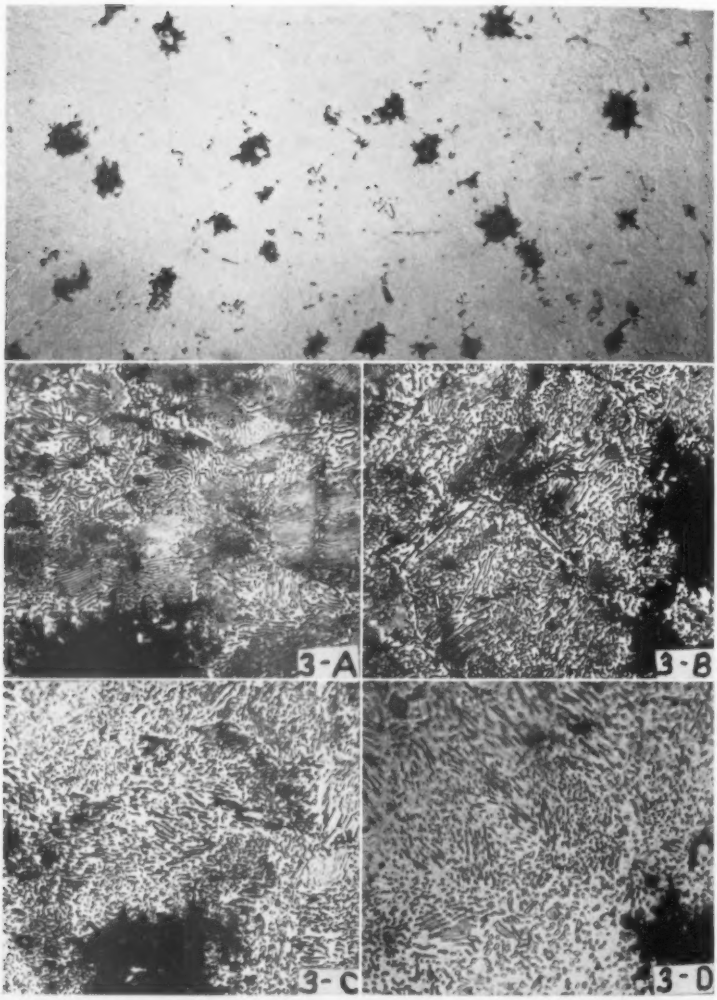

FIG. 4—SET 3. PHOTOMICROGRAPHS OF A 1.28 PER CENT MANGANESE TEST HEAT TREATED AT 1300°F. FOR VARIOUS LENGTHS OF TIME. (TOP) HEATED FOR 45 MIN. UNETCHED. X100. (CENTER LEFT) 3-A HEATED FOR 45 MIN.; 3-B HEATED FOR 3 HR.; 3-C HEATED FOR 12 HR.; AND 3-D HEATED FOR 48 HR. ETCHED IN 3 PER CENT NITAL. X500.

28. Close examination of the microstructure of Figs. 3 and 4 (sets 2 and 3) after 45 min. at 1300°F., discloses some cementite not in association with pearlite. The authors feel that the first few hours of the treatment, at 1300°F., were consumed in eliminating this pro-eutectoid cementite before graphitization of the pearlite proceeded, although it is difficult to say with assurance that the pearlite and the free cementite did not graphitize simultaneously.

29. Progress of graphitization is shown in Fig. 5, which shows the amount of combined carbon still remaining in the specimens after treatment for various periods of time.

30. The time periods were selected so that each succeeding group of bars were held at temperature four times as long as the preceding group. This permits convenient plotting of combined carbon content against a logarithmic time scale, which is shown in Fig. 6. Before the exact shape of the curves can be determined, many more points on the curves should be located with accuracy. However, it would appear that after an initial period of from 6 to 12 hr., the reduction in combined carbon proceeds linearly with the logarithm of the time of heat treatment. The slope of the curves for varying percentages of manganese may be substantially the same.

31. The fact that reduction of combined carbon does not proceed at the logarithmic rate, until several hours after heat treatment starts, may be accounted for by the fact that a certain amount of time is required to establish a carbon gradient through the ferrite, or through the grain boundaries, from the cementite in pearlite to the graphite nodules.

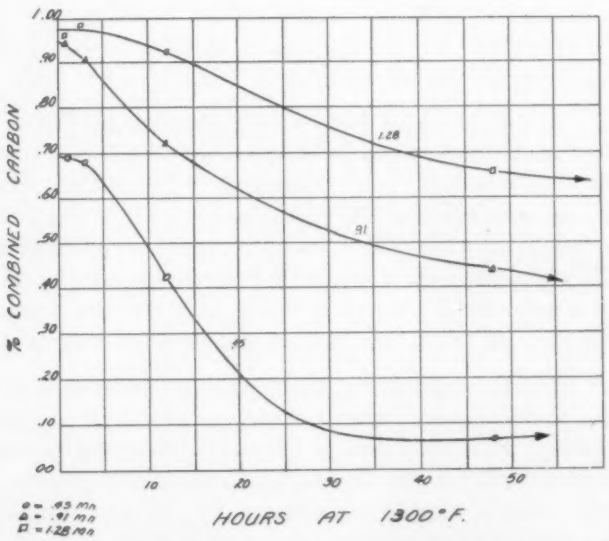

FIG. 5—GRAPH ILLUSTRATING THE AMOUNT OF COMBINED CARBON REMAINING IN SPECIMENS AFTER TREATMENT FOR VARIOUS PERIODS OF TIME.

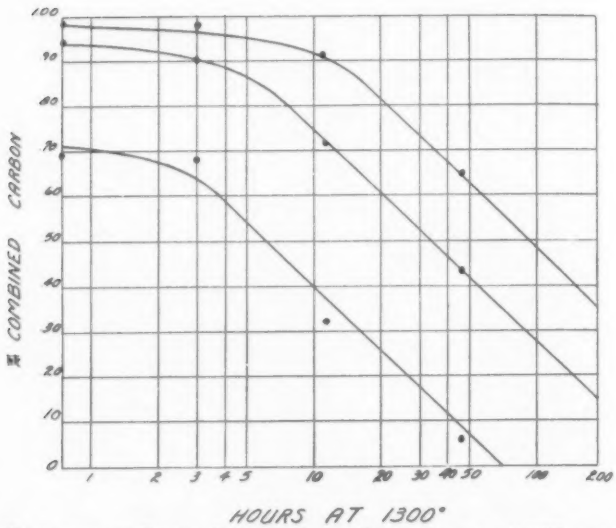

FIG. 6—VARIATIONS IN CARBON CONTENT AFTER BEING HELD AT 1300°F. FOR VARYING LENGTHS OF TIME.

TENSILE PROPERTIES

32. An examination of Fig. 7, which is drawn from the data in Table 3, indicates that at each time interval the higher manganese content material has higher strengths and lower ductility. This is, of course, due to the higher combined carbon contents resulting from the higher manganese.

33. It will be noted from an examination of Fig. 6 that a decrease in combined carbon by a decrement of 0.10 per cent, from say 0.60 to 0.50 occurs, in approximately 4 hr. in the presence of 0.45 per cent manganese. It probably occurs in from 2 to 3 hr. in iron containing the 0.25 to 0.30 per cent manganese, usually found in malleable iron. Using 0.90 per cent manganese, the same reduction requires 11 hr.; and with 1.28 per cent manganese, approximately 56 hr. is required, as estimated by exterpolation.

34. In large commercial heat treating furnaces, it is rather difficult to have every casting reach temperature at the same time and as a result, the time of holding at temperature cannot be exactly controlled for the entire load. Variation in product is apt to result. This is further complicated by the fact that ordinary variations in the other constituents, especially silicon, exert a very great influence upon graphitization rate.

35. Therefore, the use of a graphitization retardant, such as manganese, while extending the time of the heat treating cycle somewhat, does permit maintenance of physical properties within a narrow range, without the necessity of heat treating castings in very small batches.

36. The ductility of pearlitic malleable is considerably improved if the pearlite is spheroidized. The best spheroidizing temperatures lie just below the critical temperature, namely in the range from 1250 to 1325°F. Effective spheroidization requires an appreciable period of time and unless manganese, or some equivalent, is present in amounts in excess of 0.45 per cent, graphitization occurs at an excessive rate during the spheroidizing operation.

CONCLUSIONS

37. Conclusions from the data of these experiments apply with reasonable certainty only to the metal used in the tests which was produced by the air-furnace process from commercial grades of pig iron, scrap and remelt, using pulverized coal as the melting fuel. Variations from this practice, and from the type of heat treatment, might very conceivably, because of the well-known complexity of the graphitization action, give somewhat different results. Our conclusions are as follows:

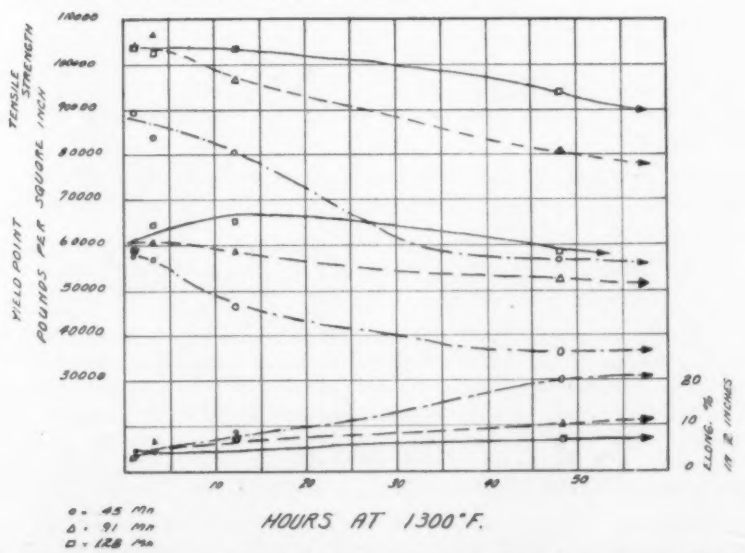

FIG. 7—DATA TAKEN FROM TABLE 3 SHOWING THAT AT EACH INTERVAL THE HIGHER MANGANESE CONTENT HAS HIGHER STRENGTHS AND LOWER DUCTILITY.

(1) The action of manganese, when used in increasing percentages in pearlitic malleable, is to retard the graphitization rate during second stage graphitization at constant temperatures below the critical.

(2) Manganese also retards the attainment of equilibrium conditions during moderately rapid cooling from graphitizing temperatures of approximately 1700°F. down to the critical temperature, in the absence of massive cementite.

(3) The time required to graphitize a definite percentage of combined carbon, at temperatures below the critical, increases rapidly as the manganese content is increased. The logarithm of the time required is, roughly proportional to the percentage of manganese increase.

(4) The principal advantages for the use of increased manganese in pearlitic malleable iron are: (a) to slow up graphitization so that in commercial operation minor variations in temperature, time, and analysis will not result in radical variations in tensile properties; (b) to prevent rapid loss of combined carbon while spheroidization of cementite is proceeding; and (c) to permit retention of combined carbon in metal subjected to a heat treating cycle which would completely graphitize ordinary malleable iron.

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2. McMillan, W. D.; *"Production of Short Cycle Malleable Iron,"* TRANSACTIONS, American Foundrymen's Association, vol. 46, pp. 697-712 (1938).

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4. Schwartz, H. A.; Johnson, H. H.; and Junge, C. H.; *"The Retarding Effect of Certain Metallic Elements on Graphitization,"* TRANSACTIONS, A.S.M., vol. 25, pp. 609-636 (1937).

DISCUSSION

Presiding: R. J. ANDERSON, Belle City Malleable Co., Racine, Wis.

MEMBER: Did the author pick this 0.45 per cent manganese iron because he presumed that would show a retarding action over the base metal of 0.35 per cent manganese?

MR. MINERT: That is right.

MEMBER: Have you any figures as to how much that retarding action was? I believe you said that it was approximately 2 hours in the case of the 0.35 per cent as compared to 4 hours in the case of the 0.45 per cent manganese.

MR. MINERT: The statement in the paper was that a decrease in combined carbon by a decrement of 0.10 per cent, from, for instance, 0.60 to 0.50, occurs in approximately 4 hours in the presence of 0.45 per cent manganese, and that it would probably occur in from 2 to 3 hours in iron containing the usual manganese, for instance in malleable iron in a range from 0.25 to 0.30 per cent manganese.

MEMBER: That was with 0.09 per cent sulphur, or approximately that, was it not?

MR. MINERT: Yes, so possibly if a person were running 0.09 per cent sulphur, they could run a little higher manganese than that.

R. J. COWAN[1]: Manganese has a very pronounced effect upon the critical point, does it not?

MR. MINERT: We have not determined any critical points. That is, we have not drawn any cooling curves on varying manganese material.

R. J. COWAN: These specimens were all annealed at a constant temperature of 1300°F. Suppose instead of selecting 1300°F. you had selected a constant temperature below the critical point? Do you suppose your result might have been more nearly comparable?

MR. MINERT: Well, 1300°F. is the temperature which we use for spheroidization in commercial practice. We have used a lower temperature; we have used 1280°F. but it does not produce as great ductility as does 1300. We do not get the elimination of as much of the combined carbon.

As for the results not being comparable, do you mean, Mr. Cowan, that the combined carbon should be nearly the same for all of the materials regardless of their manganese?

R. J. COWAN: No, there is a difference. You showed in Fig. 2 differences between the 0.45, the 0.91 and the 1.2 per cent manganese. There is a difference in the critical points. Those were all annealed at 1300°F. To be strictly comparable, ought they not be annealed at a constant difference between the critical point and the annealing temperature? That is to say, if you anneal the 0.45 per cent manganese at 1300°F., then you ought not anneal the 0.91 at 1300°F. but maybe 1325°F. or some other temperature, because the critical point has changed. Instead of using one constant annealing temperature, using a constant difference between the critical point and the annealing temperature, would not your results have come closer?

[1] Research Metallurgist, Surface Combustion Corp., Toledo, O.

MR. MINERT: I believe the point is well taken. I had not thought about it. It would be necessary, if one were to do such a thing, to determine the critical points accurately for each material. The critical point is influenced by other conditions such as the history of the metal, as has been explained in the paper. It would be necessary to use some of the metal cast for the particular experiment to determine the critical point.

C. F. JOSEPH[2]: We have made some test runs on iron, with various amounts of manganese. Our regular manganese runs about 0.38 per cent. We have made iron of 0.55, 0.75, 0.95, 1.05 per cent manganese and so on up and we find the very same thing that the author mentioned. The higher the manganese in the iron, the longer it takes to reduce the combined carbon. We have not found where there is any commercial advantage to going high in manganese. In making pearlitic malleable, a higher tensile strength is secured with the higher manganese, and possibly as time goes on, the customer might want a higher tensile iron. The commercial iron, as we are making it today, out of pearlitic malleable, seems to have ample tensile properties and therefore we have not gone to any higher manganese iron than the normal iron.

On some test runs that we have made, we found that the machinability of the higher manganese iron is adversely affected, which is a stumbling block in trying to have the customer use it. Most automobile plants look to machinability as one of their major operations. They do not like an iron which does not machine well or that machines very much harder. It is true that in the higher manganese irons, judging from what tests we have run, even with the same combined carbon resulting in the iron after annealing, or, as we call it, drawing back, the machinability is adversely affected. That is one reason why we have not pushed the iron with the higher manganese, even though the physical properties might be somewhat higher.

MEMBER: The author mentioned the fact that from a relatively rapid cool, this cementite formed in the grain boundaries. I wonder whether he would go into more detail and explain whether this relatively rapid cool was a withdrawal of the casting from the furnace or possibly a slower furnace cool.

MR. MINERT: This rapid cooling was a withdrawal of the casting from the furnace, taking it to a place where there was a good circulation of air and where it cooled very rapidly, and the structure observed was not expected at all. Commercially, the heat treatment of spheroidized pearlitic malleable with manganese used as a graphitization retardant is done by withdrawing a furnace car from the furnace, allowing the material to cool on the car to probably at least below the critical temperature—much of the material on the outside of the load in heavy loads will cool lower than 1100° F.—and then returning the load to the

[2] Metallurgist, Saginaw Malleable Iron Div., Saginaw, Mich.

furnace and reheating to the spheroidizing temperature after cooling. This cooling of the specimens was something observed just in the experimental line of treating the material. It was not anything one would run into commercially, probably.

MEMBER: Do you run the car out at 1720°F.?

MR. MINERT: Yes. Within the past few months, we have been experimenting with the effect of different treatments upon the ductility of the material and we have tried lowering the temperature of the material to about 1550°F. in the furnace, that is, just shutting off the heat and allowing the furnace to cool to 1550°F. at the rate of possibly 10 or 12 degrees per hour, and then removing it from the furnace. It gives the material a lower combined carbon and quite a bit more ductility, especially ductility under impact as observed in breaking off test lugs or in breaking up a casting and not essentially ductility such as you would observe in pulling a test bar.

MEMBER: The speaker referred to the effect of manganese above the critical point. Have you any information on how high it has to be before it takes effect?

MR. MINERT: Well, in a paper presented two years ago,* Mr. Forbes used material which had manganese varying from 0.70 to 1.00 per cent and that material did not seem to be affected appreciably by the variation in manganese as to the graphitization at the high temperature. Within that range, there did not seem to be an appreciable difference in the graphitizing properties of the material at the high temperature.

W. D. MCMILLAN[3]: This discussion is evidently based on a silicon content of about 1.00 per cent. With regard to a higher silicon iron, we have satisfactorily fully annealed a higher silicon iron with an 0.55 per cent manganese content in the same cycle which we operate to anneal an iron with a manganese from 0.30 to 0.35 per cent. With a higher silicon iron, therefore, you can stand a little more manganese. We are able to get a full anneal with 0.55 per cent manganese when the silicon and other elements are in proper relationship.

(Authors' Written Closure): Referring to Mr. Joseph's comments, it is true that it is not necessary to go to high manganese contents in order to obtain high strength pearlitic malleable iron. An examination of the chart in Fig. 4 shows that the 0.45 per cent manganese iron (and, therefore, probably standard malleable analysis of 0.35 per cent manganese or less) develops tensile strengths of the order of 90,000 lb. per sq. in. after 45 min. at 1300°F., whereas the 0.91 and 1.28 per cent manganese materials stay above 90,000 lb. per sq. in. for 27 and 55 hr. respectively. The addition of manganese serves two purposes: (1) to reduce decomposition rate of carbide and permit greater uniformity of product from large scale commercial operations, and (2) to permit

[3] Metallurgist, International Harvester Co., Chicago, Ill.

* Forbes, D. P., *"Spheroidized Pearlitic Malleable,"* TRANSACTIONS, American Foundrymen's Association, vol. 46, p. 491-512 (1938).

spheroidization to occur with improvement in ductility. This latter feature is illustrated by comparing the 0.45 per cent manganese material after ¾ hr. with the 0.91 per cent manganese material after 12 hr. The former has 89,250 lb. per sq. in. tensile strength with 4.5 per cent elongation, while the latter has 96,150 lb. per sq. in. tensile with 7.0 per cent elongation. Presumably Mr. Joseph's observation of poorer machinability with increasing manganese did not take into consideration the necessity of a longer heat-treatment after his air quench.

Mr. McMillian has pointed out that with higher silicon, a 0.55 per cent manganese can be fully annealed with a normal malleable cycle. It is difficult at this stage of the art to prophesy with any exactitude the effect of varying an analysis outside of commercial limits. Sulphur has been mentioned. This element is popularly believed to make graphitization more difficult as its percentage increases. However, assuming that substantially all the sulphur in white iron is combined with manganese in the form of manganese sulphide, then it would appear that by increasing total sulphur and keeping total manganese constant, the excess of uncombined manganese would be reduced and graphitization rates during second stage anneal would be accelerated—the exact opposite of the popular belief.

Influence of the Mold on Shrinkage in Ferrous Castings

By H. L. Womochel* and C. C. Sigerfoos*, East Lansing, Mich.

Abstract

During an investigation of the effect of mold conditions on the microstructure of gray iron, the authors observed a marked difference in the piping tendencies of the same metal poured into different molds. This observation led to studies the results of which are given in this paper. However, this paper does not record the complete investigation but is a progress report. The authors thus far have studied the effect of such properties as mold hardness, green compressive strength, permeability, moisture and sea coal on piping tendencies of 1.2 x 8-in. test bars poured in green sand molds, as well as the relative piping tendencies in dried, core-sand molds. The results obtained in this investigation then were applied to castings 6-in. in diameter, 4½-in. high, with risers 2⅛-in. in diameter and 6-in. high. In the larger type castings, the authors also studied the effect of used sand vs. new sand and natural vs. blended sands on piping properties. Such variables as metal composition and pouring temperature also were investigated. After considerable laboratory experiments, the effect of the different variables mentioned above were investigated in steel castings of the same general shape as those used in the second part of the investigation and poured from both plain carbon and low alloy steels. In addition to the experiments enumerated, the authors also have studied the effect of solidification rate, the effect of mold materials on solidification and the change in weight and dimension of test pieces with degree of piping. They advance a theory for their results which involves the expansion or contraction of the mold walls. According to their findings thus far, piping tendencies, internal shrinks and sinking of the cope face may be correlated with dimensional changes in the mold cavity; high moisture, high green compressive strength, soft ramming

* Michigan Engineering Experiment Station.

Note: This paper was presented at a Foundry Sand Research Session at the 44th A.F.A. Convention, May 8, 1940, Chicago, Ill.

and unfavorable grain distribution of the sand tend to promote shrinkage defects in gray iron; sea coal decreases the amount of piping in gray iron poured in green sand; small gray iron castings made in baked core sand molds show no shrinkage defects; and steel castings are less susceptible to influence of the mold than gray iron castings.

1. In conducting an investigation to determine the influence of mold conditions on the microstructure of cast iron, a number of 8 x 1.2-in. bars were cast from the same ladle in open molds of various types. It was observed that a marked difference existed in the piping tendency in the molds.

2. Fig. 1 shows cross sections of the tops of three of the bars. The casting labeled "green" was poured in a synthetic green sand, and the one marked "core" in an oil sand core. The dry mold was prepared from silica sand, bentonite, and a small percentage of a plastic refractory, and was ignited at 1600°F. to remove moisture and organic material.

3. These results indicated the possibility of some relation between mold material and shrinkage defects in castings made in closed molds. A number of castings were made from several patterns in green sand and core sand to determine the extent of the influence of the mold. Fig. 2 illustrates cross sections of three castings poured from the same ladle in the order shown. The dimension of the casting at the widest point is 2¼-in. Castings 1 and 3 were made in two types of core sand; casting 2, in green sand from the foundry floor. The molds were horn gated and no risers were used.

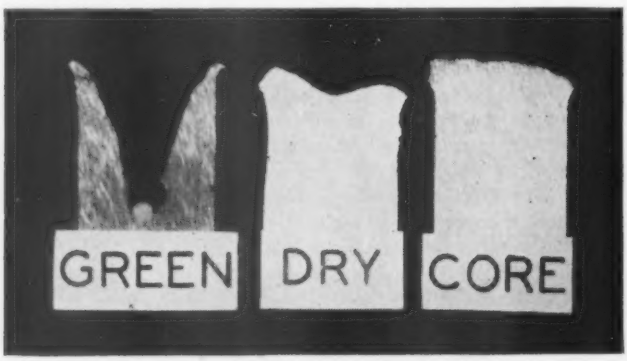

FIG. 1—EFFECT OF MOLD ON PIPING TENDENCY.

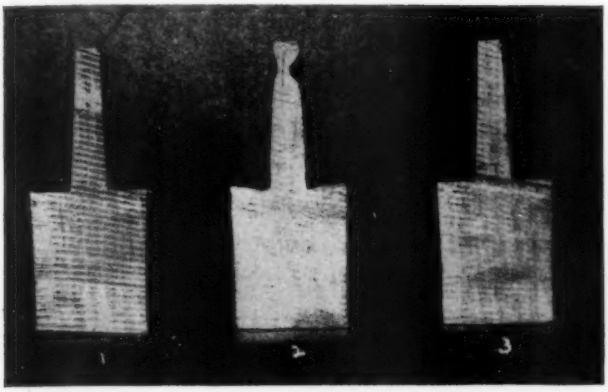

FIG. 2—SECTIONS OF CASTINGS FROM CORE AND GREEN SAND MOLDS.

4. Casting 2 (Fig. 2) shows a sinking of the cope face of the heavy section, and a constriction of the light section. The machined surfaces of the castings were ground smooth and subjected to a prolonged etch. No internal porosity was revealed.

5. The difference in shrinking tendency between the dry and green molds suggested the likelihood of some influence of the various kinds and conditions of green sand on shrinkage defects. A number of experiments have been conducted to investigate this possibility.

RESULTS

6. The castings made in the preliminary tests showed a correlation between the small bars cast in open molds and castings poured through a gate in a closed mold. It was decided to employ the top-poured bar as a convenient test casting for experiments to determine the direction of the influence of such factors as mold hardness, permeability, moisture content, and green compressive strength.

7. The gray iron for these experiments was melted in an indirect arc electric furnace. Metal compositions for the series of vertical bar castings were within the following limits: Carbon, 2.75-3.00 per cent; silicon, 2.15-2.40 per cent; manganese, 0.70-0.90 per cent; phosphorus, 0.10-0.20 per cent; sulphur, 0.07-0.10 per cent. Charges consisted of returns, pig, steel, and malleable scrap. A late addition of 0.25 per cent or more of silicon as 75 per cent ferrosilicon was made to the furnace immediately before tapping.

Tapping temperatures were in the range 2750 to 2800°F. Pouring temperatures were in the range 2575 to 2625°F., as indicated by an optical pyrometer. Carbon content was regulated by graphite addition after the meltdown. Examination of sections from the 1.2 x 8-in. bars showed a normal structure in most cases.

8. Most of the sands used in these experiments were prepared by mulling mixtures of lake sand, bank sand, and a sand containing 28 per cent clay.

9. All of the castings shown in any one illustration were poured from the same ladle, unless otherwise indicated. Variations in metal temperature, resulting from cooling in the ladle during pouring, were minimized by pouring as rapidly as possible. In checking these experiments, the order of pouring was reversed with no apparent effect on the results obtained.

Influence of Mold Hardness

10. Fig. 3 shows the influence of mold hardness. Casting 1 was poured in a core sand mold. Casting 2 was cast in a green sand mold, rammed to an average hardness of 90 as determined by a mold hardness tester. Castings 3 and 4 were poured in molds of the same type rammed to a mold hardness of 50 and 5 respectively. Casting 4 was swollen at the lower end.

11. These results show the relation between piping and swelling. In the case of a very soft mold, swelling evidently continues for some time after pouring has ceased and the casting has partially solidified. The volume occupied by swelling is fed by piping or "shrinkage." This experiment was checked using cylinders of 2-in. diameter with identical results.

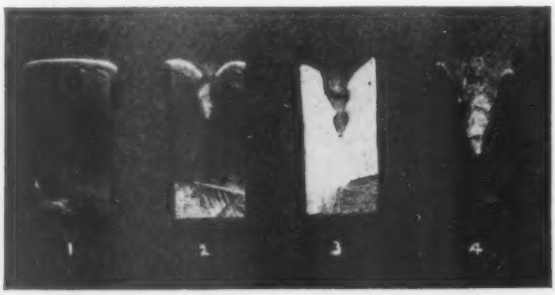

FIG. 3—EFFECT OF MOLD HARDNESS ON PIPING TENDENCY.

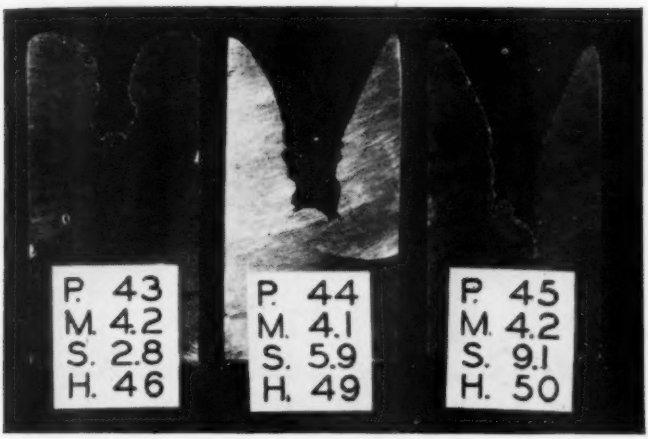

FIG. 4—EFFECT OF GREEN COMPRESSIVE STRENGTH ON PIPING TENDENCY.

Effect of Green Compressive Strength

12. Fig. 4 shows the effect of green compressive strength. The castings shown are the top sections of bars 1½-in. in diameter and 12-in. long. In this and subsequent illustrations, *P* refers to permeability, *M* to moisture content, *S* to green compressive strength, and *H* to mold hardness.

13. A base sand, consisting of a mulled mixture of the previously mentioned clay containing sand, bank, and lake sands was employed. Green compressive strength was raised from 2.8 to 9.1 by additions of bentonite. Moisture and mold hardness were kept as nearly constant as possible. These castings were molded by squeezing a plate in a 3½-in. cope and drag. The molds were poured on end, through a hole in the flask, at the parting line.

14. This experiment indicates that, rammed to equal hardness, the low green strength sands have less tendency to cause piping than high strength sands when the increase in strength is produced by additions of rich clay. The hardness values are the average of a number of readings taken at fixed points on the parting surface.

15. The values for permeability and green compressive strength were obtained from the standard laboratory tests in all of these experiments, and do not refer to mold values.

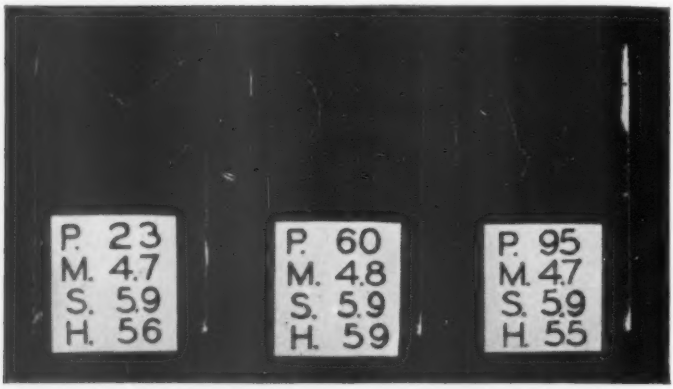

FIG. 5—INFLUENCE OF PERMEABILITY ON PIPING TENDENCY.

Influence of Permeability

16. Fig. 5 shows the effect of permeability. A base sand, consisting of bank sand and the clay-containing sand, was used. Permeability was increased from 23 to 95 by substituting lake sand for the bank sand in increasing proportions. Molding was conducted in the manner described for the previous experiment. Piping tendency is increased with permeability when permeability is altered by changing the grain size of the sand.

Influence of Moisture

17. Fig. 6 shows the influence of moisture. Sand was taken from a heap on the foundry floor and the moisture varied from

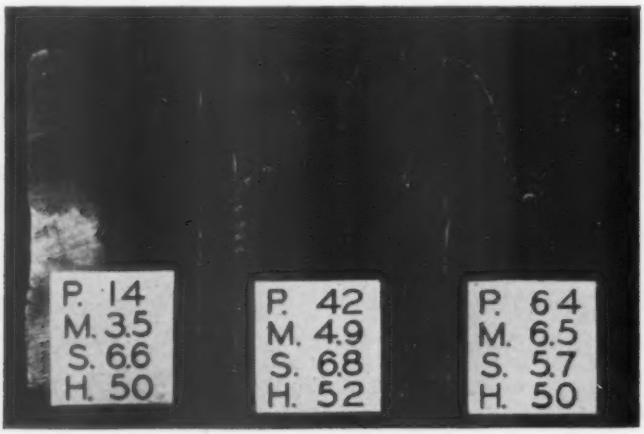

FIG. 6—EFFECT OF MOISTURE ON PIPING TENDENCY.

3.5 to 6.5 per cent. Permeability increased with the moisture. This sand and the sands of Figs. 1 to 5 contained no sea coal.

Influence of Sea Coal

18. Fig. 7 illustrates the effect of sea coal. The letter *C* on the illustration refers to per cent sea coal by weight. A mixture of heap sand, clay-containing sand, bank sand, and lake sand was used as a base. Permeability and green compressive strength were kept constant by varying the proportions of the constituents. This experiment shows that sea coal has a remarkable effect on the piping tendency when it is added to a sand and permeability and green strength maintained.

RESULTS APPLIED TO HEAVY CASTINGS

19. To determine the extent to which the results of Figs. 3 to 7 could be applied to heavy castings made in closed molds, a series of experiments were conducted using the casting shown in Fig. 8. Pouring temperature for the heavy castings was 2575—2625°F. unless otherwise indicated.

20. Molds for the heavy castings were made by hand ramming followed by squeezing. The gated pattern employed was mounted on a match plate. During these experiments, a fillet was added to the pattern at the junction of the riser and casting. This fillet does not appear on all of the castings in the photographs.

21. With the exception of the experiments shown in Figs. 19, 20, and 21, the heavy castings were cast in sands prepared by mull-

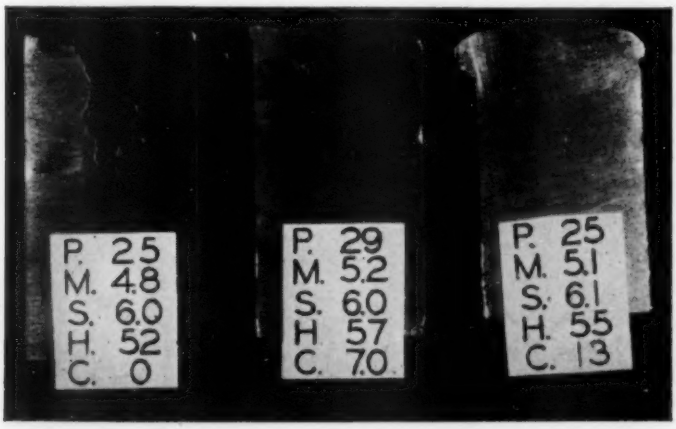

FIG. 7—RELATION BETWEEN SEA COAL CONTENT AND PIPING TENDENCY.

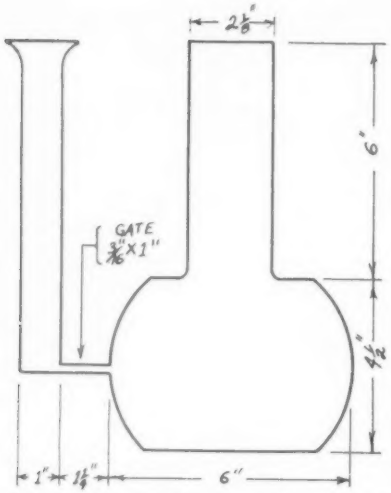

FIG. 8—HEAVY SECTION CASTING FOR SHRINKAGE TESTS.

ing mixtures of lake, bank, and clay-containing sands as previously described.

Core Sand vs. Green Sand

22. Fig. 9 shows a comparison of core sand with green sand, and with green sand plus sea coal. Metal composition: Carbon, 2.96 per cent; silicon, 2.32 per cent; manganese, 0.78 per cent; phosphorus, 0.18 per cent; sulphur, 0.072 per cent. In this and

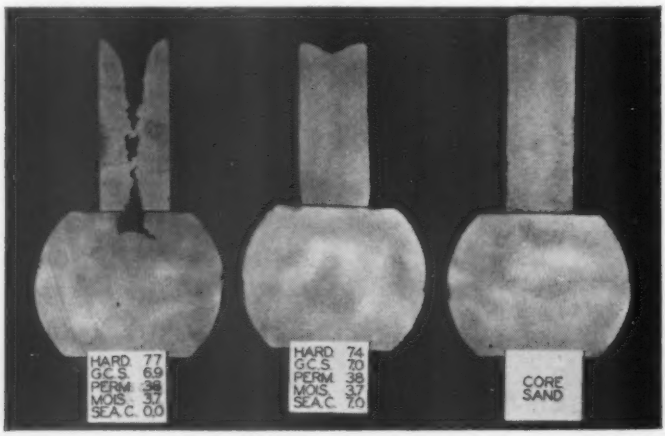

FIG. 9—CORE SAND COMPARED WITH GREEN SANDS.

subsequent figures, *"Hard"* refers to mold hardness determined by averaging a number of readings taken at fixed points on the mold surface. *"G. C. S."* indicates green compressive strength; *"Perm,"* permeability; *"Mois,"* moisture; and *"Bent,"* bentonite. The results of this experiment are in agreement with those of Figs. 2, 3, and 7.

23. In a separate experiment (not illustrated) the composition of the core sand was varied through wide limits of oil and cereal content with no appreciable effect on piping or soundness of the castings.

24. While conducting this investigation, a number of molds were sectioned and hardness readings taken at various points beneath the surface. These tests indicated that readings taken on the mold surface were representative of the hardness throughout the mold.

Influence of Green Compressive Strength

25. Fig. 10 shows the influence of green compressive strength. A base sand of bank, lake, and clay-containing sands was employed. Green compressive strength was increased from 4.3 to 10.2 lb. per sq. in. by additions of one and 3 per cent bentonite. Results agree with those of Fig. 4. Addition of the bentonite produced an appreciable increase in permeability. Metal composition: Carbon, 2.93 per cent; silicon, 2.11 per cent; manganese, 0.77 per cent; phosphorus, 0.15 per cent; sulphur, 0.064 per cent.

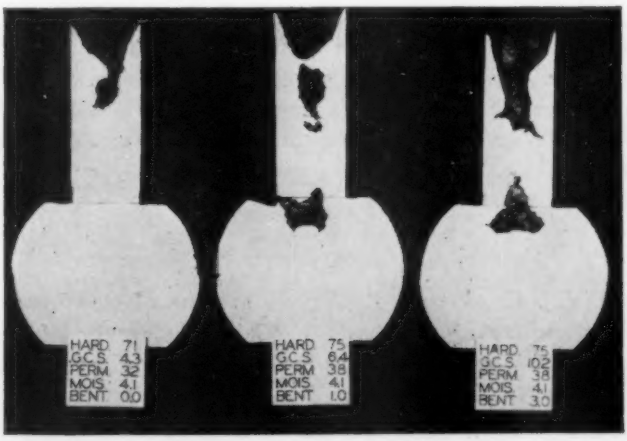

FIG. 10—EFFECT OF GREEN COMPRESSIVE STRENGTH.

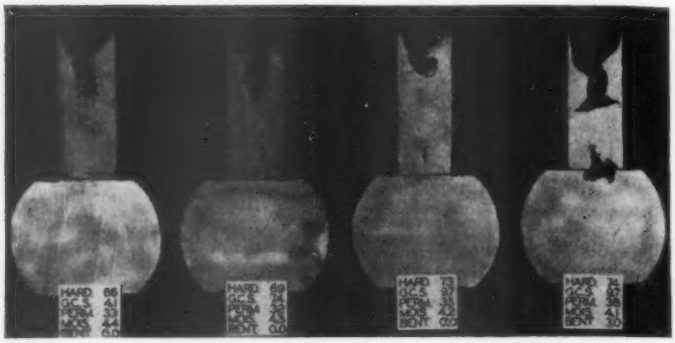

FIG. 11—EFFECT OF GREEN COMPRESSIVE STRENGTH.

26. Fig. 11 also shows the influence of green compressive strength. The sands for castings 1, 2, and 3 (left to right) were prepared by mulling mixtures of the clay-containing, lake, and bank sands with increasing proportions of the clay-containing sand to raise the strength from 4.1 to 9.7 lb. per sq. in. Permeability was maintained by varying the proportions of lake and bank sands. The sand for casting 4 was the same as that for casting 1, except that 3 per cent bentonite was added. Comparison of castings 1 and 4 shows good agreement with Fig. 10. The difference between castings 3 and 4 would indicate that the manner of adding clay has an effect on the results.

27. It should be noted, however, that the distribution of grain size is not controlled in this experiment. The presence of fines or pan material in sand doubtless affects the distribution of clay and moisture, and may have an influence on piping tendency. The increase in the amount of clay-containing sand probably altered the grain size distribution to a much greater extent than did the addition of 3 per cent bentonite. Metal composition: Carbon, 3.11 per cent; silicon, 2.26 per cent; manganese, 0.83 per cent; phosphorus, 0.13 per cent; sulphur, 0.068 per cent.

Influence of Type of Clay Bond

28. Fig. 12 shows the influence of type of clay bond. This experiment was performed as a check on the results of Fig. 11. The low green strength sand was prepared by mulling bank, lake, and the clay-containing sands. For casting 1 (left to right), the strength was increased to 9.5 lb. per sq. in. by the addition of 2.5 per cent bentonite. The sand for casting 3 was prepared by increas-

FIG. 12—EFFECT OF TYPE OF CLAY BOND ON PIPING TENDENCY.

ing the amount of clay-containing sand. Permeability was maintained by changing the proportion of lake sand in the mixture. Metal composition: Carbon, 3.16 per cent; silicon, 1.63 per cent; manganese, 0.75 per cent; phosphorus, 0.12 per cent; sulphur, 0.075 per cent.

29. This experiment was repeated a number of times with good agreement, except in one instance. In this case, the high clay-containing sand mixture showed a defect at the base of the riser, similar to that found in the bentonite sands.

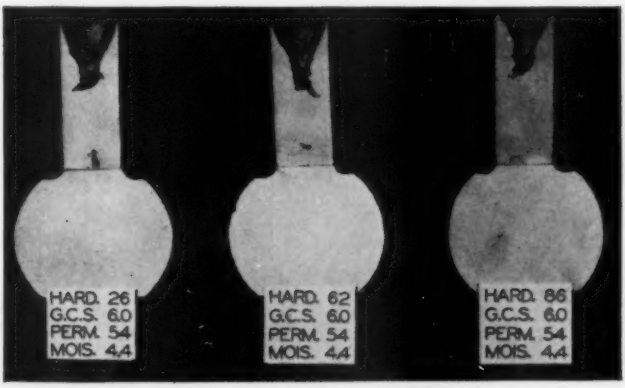

FIG. 13—EFFECT OF MOLD HARDNESS.

Influence of Mold Hardness

30. Fig. 13 shows the influence of mold hardness. Hardness was varied from 26 to 86 by increasing the hardness of ramming and squeezing pressure. The values for green compressive strength and permeability were obtained from the standard laboratory tests and do not refer to the mold values. The influence of mold hardness over the range tested appears to be small. A slight tendency towards internal shrinks is evidenced by the porosity near the base of the risers of the two softer molds. Metal composition: Carbon, 3.36 per cent; silicon, 1.53 per cent; manganese, 0.85 per cent; phosphorus, 0.15 per cent; sulphur, 0.068 per cent.

Effect of Permeability

31. Fig. 14 shows the effect of permeability. Permeability of the sands was varied by changing the proportions of lake, and bank sands to give the grain fineness numbers indicated. The sands were prepared by mulling the constitutents in the following proportions:

	Sand 1	*Sand 2*	*Sand 3*
Clay-Containing, per cent	37	37	37
Bank, per cent	63	8	
Lake, per cent		55	63

The following data was obtained from the sand analysis:

	Sand 1	*Sand 2*	*Sand 3*
A. F. A. Fineness No.	115	69.1	61.4
Clay Content, per cent	10.1	10.6	10.6
Fines, per cent	36.6	9.4	7.4

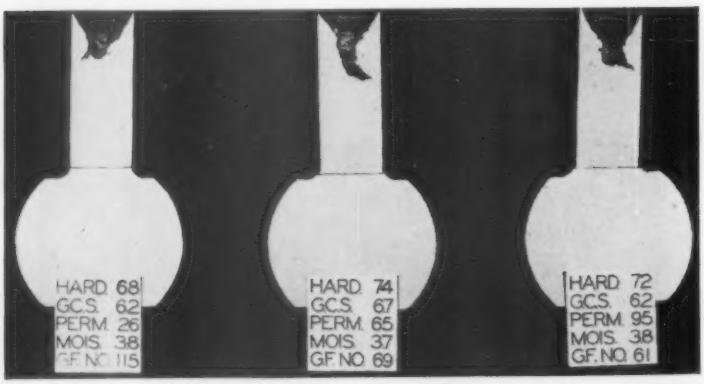

FIG. 14—EFFECT OF PERMEABILITY.

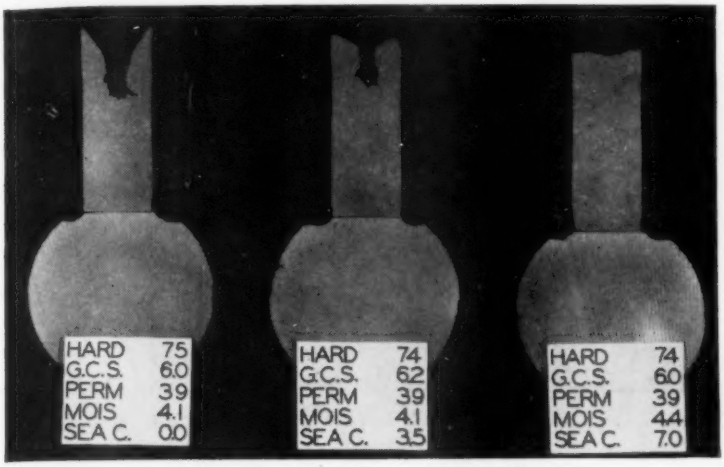

FIG. 15—EFFECT OF SEA COAL ON PIPING TENDENCY.

32. Metal composition: Carbon, 3.45 per cent; silicon, 1.51 per cent; manganese, 0.65 per cent; phosphorus, 0.18 per cent; sulphur, 0.082 per cent. Decreasing the grain fineness number, while holding the clay content constant, produced some increase in piping tendency.

Effect of Sea Coal

33. Fig. 15 indicates the effect of sea coal. Sea coal is expressed in per cent by weight. The addition of sea coal produces a marked decrease in piping tendency. In the case of the high sea-coal sands, some feeding of the casting apparently occurs from constriction of the riser. This is apparent in casting No. 3 at the junction of the casting and riser. Metal composition: Carbon, 3.25 per cent; silicon, 1.49 per cent; manganese, 0.69 per cent; phosphorus, 0.11 per cent; sulphur, 0.095 per cent.

Effect of High Strength Plus Sea Coal

34. Fig. 16 shows the combined effect of high strength and sea coal. Sand for casting 2 (left to right) was prepared by adding 7 per cent sea coal by weight to the base sand used for casting 1. Sand for casting 3 was prepared by adding 2 per cent bentonite to the base sand. Sand 4 is the base sand plus bentonite and sea coal. Comparison of castings 2 and 4 shows that high strength is effective in increasing shrinkage in the presence of sea coal. Constriction of the riser is particularly noticeable in the case of cast-

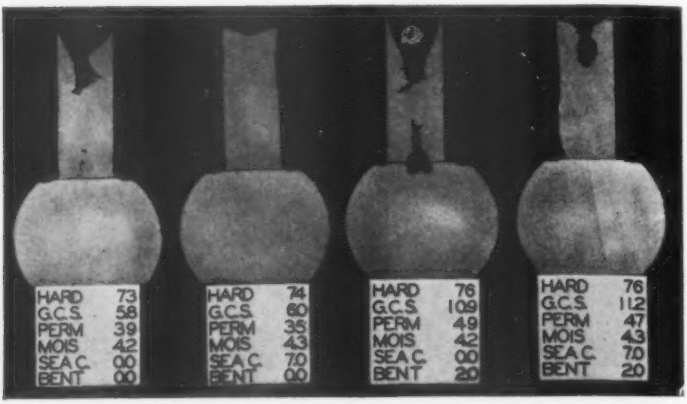

FIG. 16—EFFECT OF HIGH STRENGTH AND SEA COAL ON PIPING TENDENCY.

ing 4. Composition of iron: Carbon, 3.25 per cent; silicon, 1.53 per cent; manganese, 0.77 per cent; phosphorus, 0.19 per cent; sulphur 0.075 per cent.

Effect of Used Sand

35. Fig. 17 compares a new and a used sand. Sand for casting No. 2 was used once before making the casting shown in the illustration. The 7 per cent refers to original sea coal content of the sand. The experiment shows that use appreciably increases the piping tendency of the sand. Analysis of iron: Carbon, 3.29 per

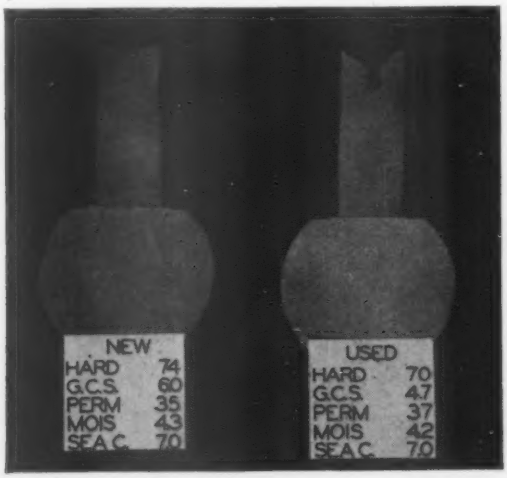

FIG. 17—INFLUENCE OF USE OF SAND ON ACTION OF SEA COAL.

FIG. 18—COMPARISON OF BLENDED AND NATURAL SANDS.

cent; silicon, 1.53 per cent; manganese, 0.77 per cent; phosphorus, 0.19 per cent; sulphur, 0.075 per cent.

Natural vs. Blended Sands

36. Fig. 18 compares blended and natural sands. A blended sand was prepared by mulling a mixture of the previously mentioned clay-containing, bank and lake sands to approximate the properties of an Albany sand. Metal composition: Carbon, 3.44 per cent; silicon, 1.49 per cent; manganese, 0.63 per cent; phosphorus, 0.15 per cent; sulphur, 0.096 per cent.

Effect of Moisture on Albany Sand

37. Fig. 19 shows the effect of moisture content on a No. 3

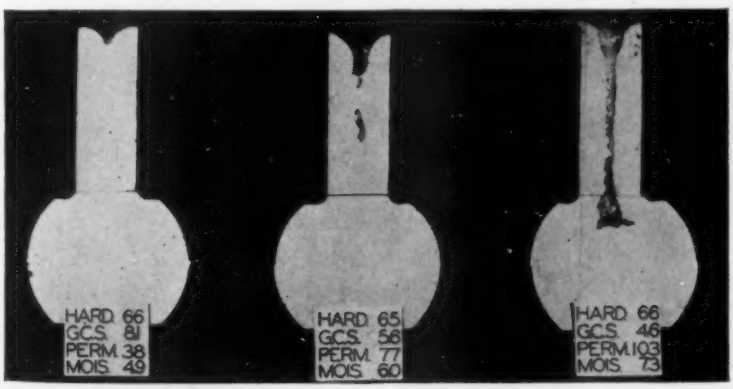

FIG. 19—EFFECT OF MOISTURE CONTENT IN NATURAL SANDS ON PIPING TENDENCY.

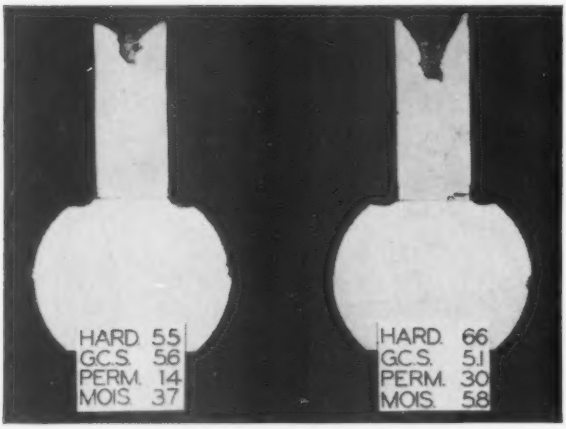

FIG. 20—INFLUENCE OF MOISTURE IN USED SANDS.

Albany sand. Increasing the moisture content from 4.9 to 7.3 per cent, increased piping tendency. Metal composition: Carbon, 3.10 per cent; silicon, 1.42 per cent; manganese, 0.63 per cent; phosphorus, 0.17 per cent; sulphur, 0.089 per cent. This experiment shows that natural sands influence the piping tendency in the same manner as the blended sands.

Influence of Moisture in Used Sands

38. Fig. 20 shows influence of moisture on piping in used sands. Most of the sands described under Figs. 1 to 19 were new. In order to determine the extent to which the results were applicable to used, natural-bonded sands, a quantity of sand was taken from a heap in a jobbing foundry and tests run to determine the effect of moisture variations on the piping tendency. Although the sand for casting 2 was rammed somewhat harder than that for casting 1, an appreciable increase in the volume of the pipe was produced when the moisture was increased from 3.7 to 5.8 per cent. As taken from the original heap, this sand contained 7.5 per cent moisture, and was giving trouble with shrinkage in small castings. This sand contained no sea coal. Metal composition: Carbon, 3.26 per cent; silicon, 1.42 per cent; manganese, 0.62 per cent; phosphorus, 0.20 per cent; sulphur, 0.079 per cent.

Effect of Green Compressive Strength in Used Sands

39. Fig. 21 shows the effect of green compressive strength on used sands. The strength was changed by clay additions to a

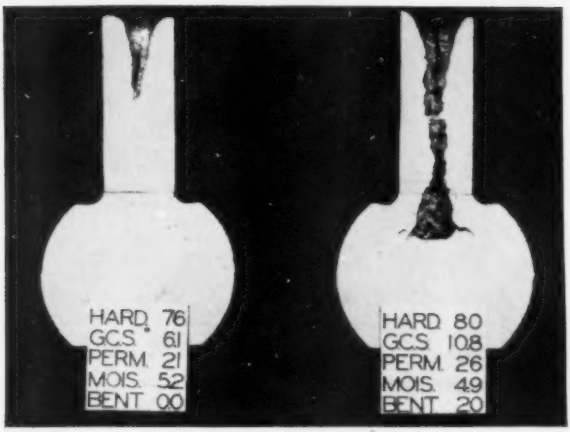

FIG. 21—EFFECT OF GREEN COMPRESSIVE STRENGTH IN USED SANDS.

used sand. Two per cent bentonite was added to sand described under Fig. 20. The results of Figs. 20 and 21 indicate that old, natural-bonded sands are affected in the same manner as new and blended sands by variations in moisture and strength. Metal analysis: Carbon, 3.11 per cent; silicon, 1.29 per cent; manganese, 0.67 per cent; phosphorus, 0.15 per cent; sulphur, 0.075 per cent.

Influence of Metal Composition

40. Fig. 22 shows the influence of metal composition. The change in analysis did not affect the shrinkage appreciably in either the core or green sand molds.

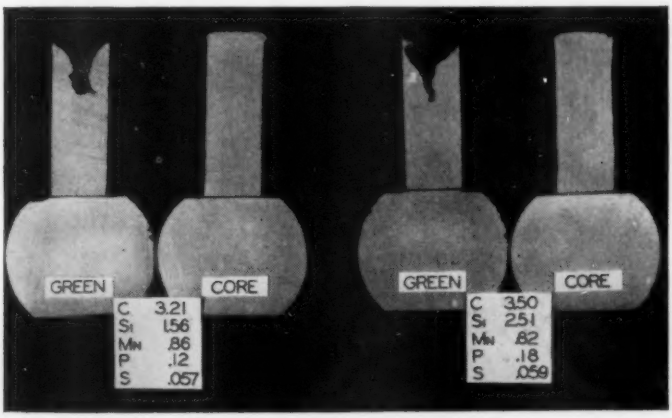

FIG. 22—EFFECT OF METAL COMPOSITION.

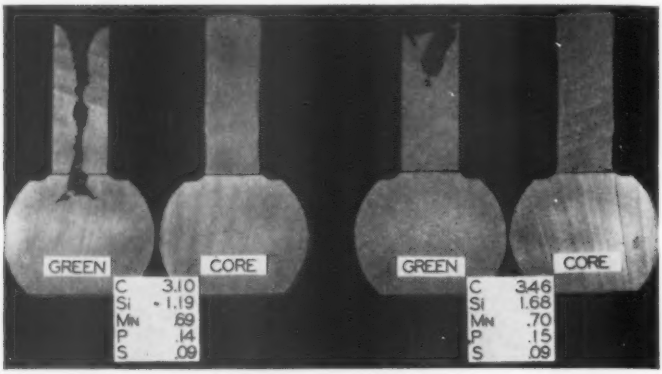

FIG. 23—EFFECT OF METAL COMPOSITION.

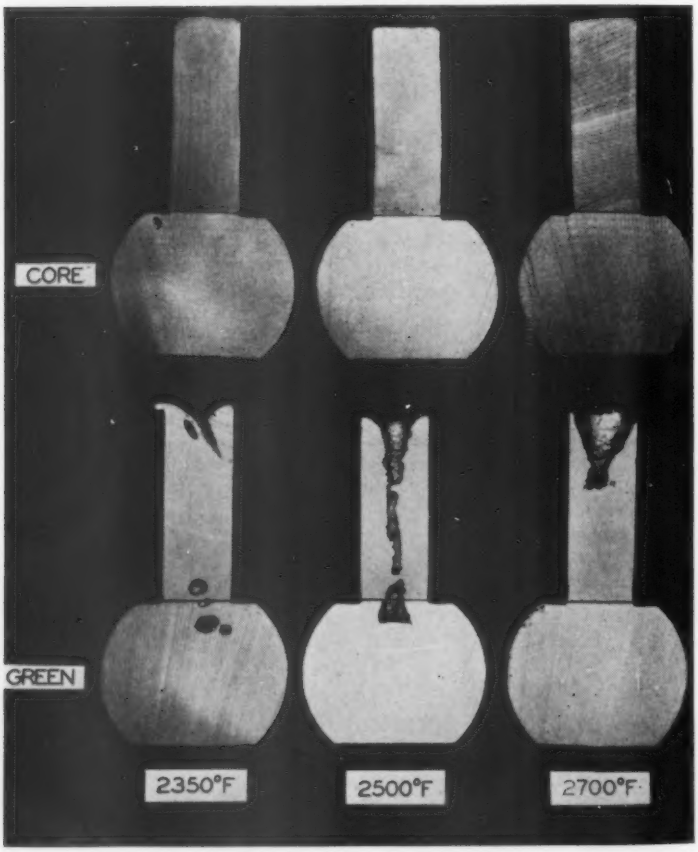

FIG. 24—EFFECT OF POURING TEMPERATURE.

41. Fig. 23 also shows the effect of metal composition. Decreasing the carbon and silicon contents produced a marked increase in shrinkage in the case of the green sand castings. It is interesting to note that the castings made in core sand appear to be unaffected by the change in metal composition.

Effect of Pouring Temperature

42. Fig. 24 illustrates the effect of pouring temperature. A core and a green sand mold were poured at each of the indicated temperatures. All of the castings were from the same ladle. Temperatures were determined with an optical pyrometer. Metal composition: Carbon, 3.05 per cent; silicon, 1.86 per cent; manganese, 0.81 per cent; phosphorus, 0.21 per cent; sulphur, 0.085 per cent.

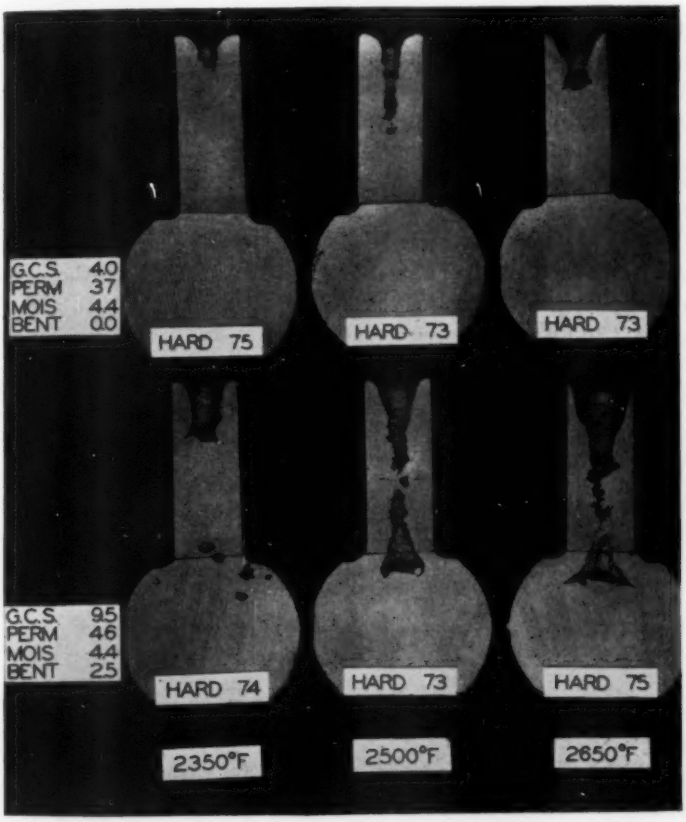

FIG. 25—INFLUENCE OF POURING TEMPERATURES.

43. Fig. 25 also shows the effect of pouring temperature. A high strength (bottom row) and a low strength (top row) sand mold was poured at each of the temperatures indicated. All of the castings were from the same ladle. Metal composition: Carbon, 3.17 per cent; silicon, 1.34 per cent; manganese, 0.71 per cent; phosphorus, 0.11 per cent; sulphur, 0.075 per cent. The results of Figs. 25 and 26 show that the influence of mold material on shrinkage is not confined to a limited temperature range. Lowering the pouring temperature from 2650 to 2500°F. appears to deepen the pipe and decrease its diameter. The apparent reproducibility of the defects in the low temperature castings of both illustrations is of interest. These defects have more the appearance of gas holes than of shrink cavities.

44. Fig. 26 shows castings poured in synthetic sand, natural sand, and core sand. These test castings were poured in a Wisconsin foundry which was having trouble with shrinkage in a moderately heavy job molded in green sand. The difficulty was overcome by casting in dry sand molds. Some experiments were performed in this foundry using these castings, which are 3-in. cylinders, as test specimens. Metal composition approximately: Carbon, 3.00 per cent; silicon, 1.90 per cent; manganese, 0.80 per cent; phosphorus, 0.20 per cent; sulphur, 0.10 per cent. Pouring temperature 2350°F. The casting marked "*syn*" was a synthetic sand. On this photograph, "*P*" refers to permeability, "*M*" to moisture, and "*S*" to green compressive strength.

EXPERIMENTS WITH STEEL CASTINGS

45. A number of experiments were performed to extend the investigation to steel castings. The results are shown in Figs. 27

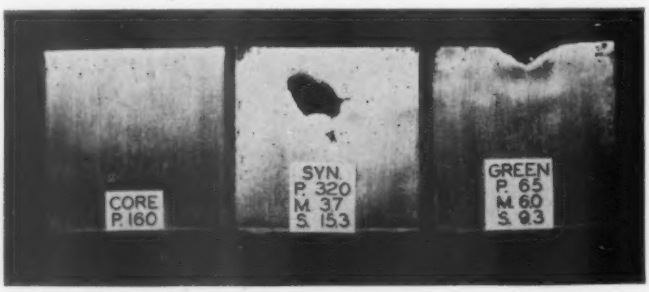

FIG. 26—SYNTHETIC SAND COMPARED WITH NATURAL AND CORE SANDS.

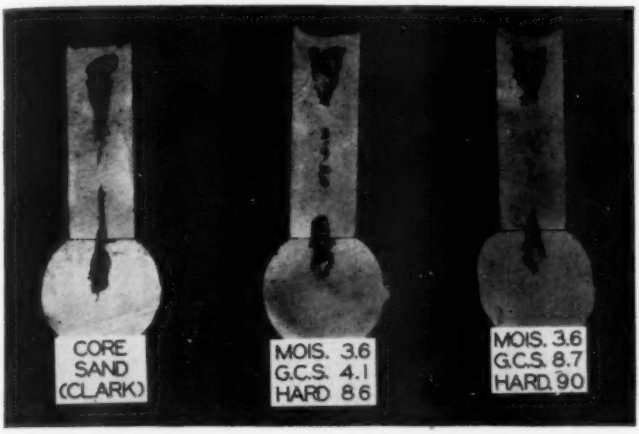

FIG. 27—INFLUENCE OF SAND ON SHRINKAGE IN STEEL.

to 30. The castings were poured in the foundry of the Clark Equipment Co., Buchanan, Michigan.

46. The steel castings were in the form of a ball of 4-in. diameter with a 2¼-in. riser. Gating was at the parting line. Pouring temperatures were in the range 2900 to 2950°F.

47. Fig. 27 shows some steel castings poured in core and green sands. The green sand was a synthetic sand prepared by mulling silica sand, bentonite, cereal, and oil. As ordinarily used, it shows a permeability in excess of 200, and a green compressive strength of 6.5 lb. per sq. in. at 3 per cent moisture. Casting No. 3 (at the right) was poured in the same sand as casting No. 2, except that the strength was raised from 4.1 to 8.7 lb. per sq. in. by increasing the amount of bentonite. Composition of steel: Carbon, 0.25-0.35 per cent; manganese, 0.65-0.80 per cent; silicon, 0.40-0.50 per cent; sulphur, 0.04 per cent; phosphorus, 0.04 per cent.

Effect of Mold Hardness

48. Fig. 28 shows the results of an experiment performed to determine the effect of variations in mold hardness. Steel composition the same as for Fig. 27.

49. Fig. 29 shows the effect of moisture on shrinkage in steel. The moisture contents are the usable extremes for the sand. The steel used in pouring these castings was a manganese-vanadium steel (manganese, 1.50-1.65 per cent; vanadium, 0.10-0.20 per cent).

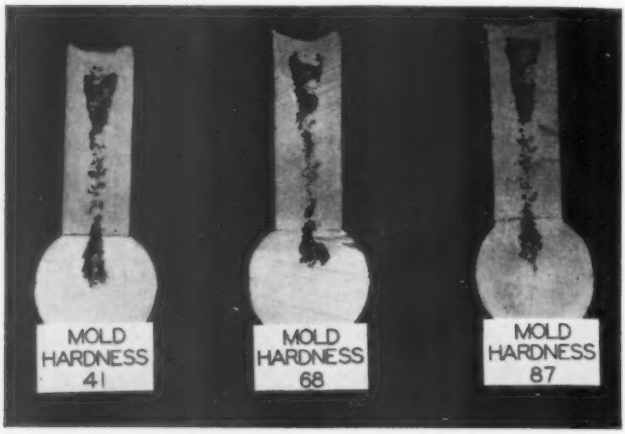

FIG. 28—EFFECT OF MOLD HARDNESS ON SHRINKAGE IN STEEL CASTINGS.

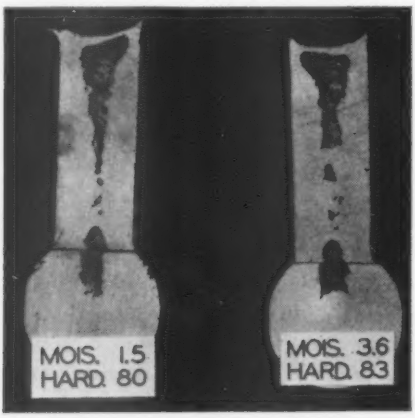

FIG. 29—EFFECT OF MOISTURE ON SHRINKAGE IN STEEL.

Effect on Type of Clay Bond

50. Fig. 30 shows the effect of changing type of clay bond. The casting marked "Wing" was poured in a sand prepared by mulling the clay-containing sand with silica sand. The second casting was poured in the synthetic sand described under Fig. 28. The effect of the lower moisture and difference in clay is small. Steel composition is the same as for Fig. 29.

DISCUSSION

Solidification Rate Experiments

51. While conducting these experiments, it seemed possible

FIG. 30—EFFECT OF TYPE OF CLAY BOND.

that cooling or solidification rate was partly responsible for the differences observed. To determine the solidification rate in green and core molds, a number of molds were made, as illustrated in Fig. 31.

52. Fig. 31 shows a sectional view of the mold used for determining the relative rate of solidification in green and core sands. Cylindrical castings of 2¾-in. diameter and 10-in. long were poured with gray iron through a common gate as shown in the diagram. After solidification was partly completed, the two copes, which were of the different sands, were lifted at the same time, and the liquid metal allowed to drain or "bleed" from the cylinders.

53. A number of these experiments were performed. In most cases, the shell of solidified metal was approximately the same

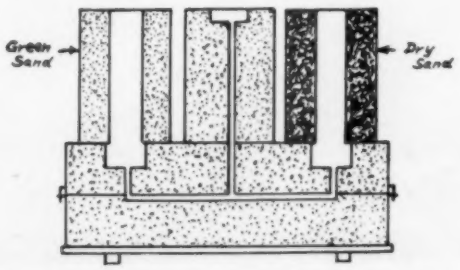

FIG. 31—MOLD FOR DETERMINING RATES OF SOLIDIFICATION IN CORE AND GREEN SANDS.

thickness in the two cylinders. Figs. 32 and 33 show cross sections of the extreme cases obtained. These experiments were confined to comparisons of green and core sand molds. Results led to the conclusion that solidification occurred at approximately the same rate in the two copes, and that the difference was too small to provide an explanation for the marked difference in piping in castings made in green and core sand.

Influence of Mold Materials on Solidification

54. As an additional check on the influence of cooling rate, a number of molds were poured with thermocouples, as illustrated

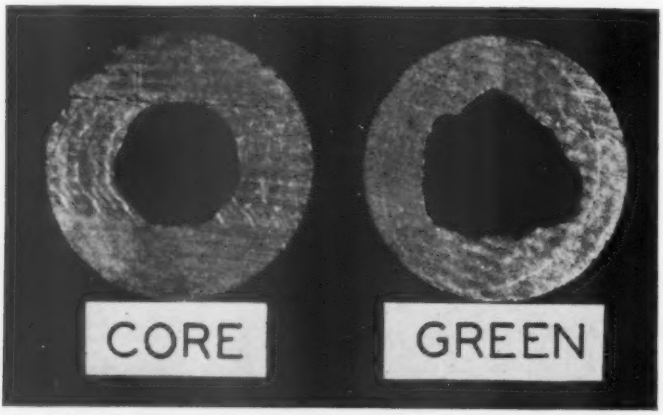

FIG. 32—CROSS SECTIONS OF CYLINDERS FROM MOLD OF FIG. 31.

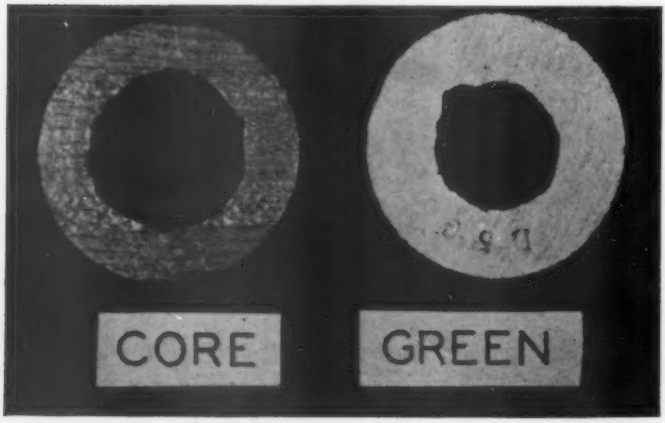

FIG. 33—CROSS SECTIONS OF CYLINDERS FROM MOLD OF FIG. 31.

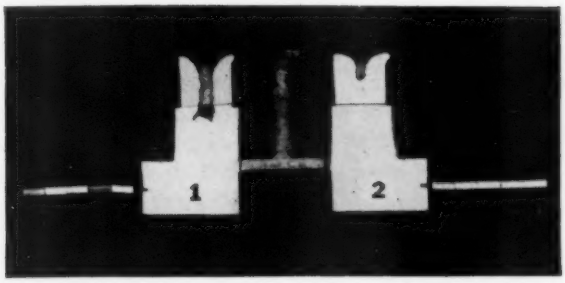

FIG. 34—ARRANGEMENT OF THERMOCOUPLES FOR DETERMINING INFLUENCE OF SAND ON COOLING RATE.

in Fig. 34. The mold for these castings was made up of two sands. In Fig. 34 casting number 1 was rammed up in a sand containing 2 per cent bentonite. Casting 2 was made in the same base sand as casting 1 without the bentonite addition. Chromel-alumel couples were inserted in the walls of the mold as shown in Fig. 34. The section size at the point of insertion of the thermocouples was 2½ x 4 x 4½-in. The castings were poured through the common gate. Pouring temperature was 2475°F.

55. The curves marked "bentonite" and "no bentonite" in

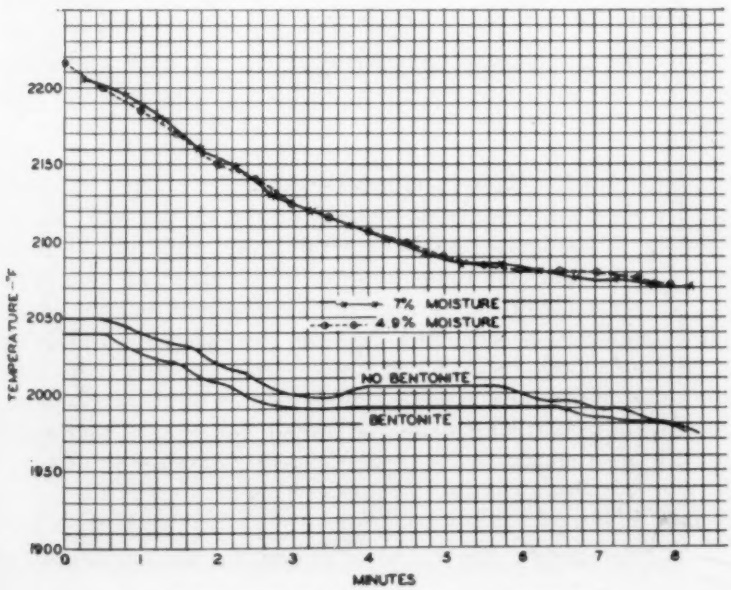

FIG. 35—INFLUENCE OF MOLD MATERIAL ON COOLING RATE.

Fig. 35 were obtained from the castings of Fig. 34. The two upper curves were from castings poured in No. 3 Albany sand with moisture contents as indicated. The higher temperatures of the moisture curves are to some extent a result of the positions of the couples in the mold. In the moisture experiment, the couple-leads were insulated with fused quartz tubes and inserted farther into the mold cavity. Temperature readings were taken as soon as possible (4 or 5 sec.) after pouring. In the case of each pair of castings poured in this manner, the shrinkage differences observed were in agreement with the previous experiments described under Figs. 1 to 31.

Micro-examination Results

56. Sections from various positions in the heavy castings have been examined under the microscope to determine the presence of any influence on microstructure of such factors as sea coal, moisture, and clay contents. At the present time, no correlation between microstructure and shrinkage or piping tendency has been detected in the heavy sections. Unless shrinkage differences can be attributed to very small changes in pearlite fineness, or graphite size and distribution, it must be concluded that changes in the character of the metal under the influences of the mold are not responsible for the differences in shrinkage.

57. In the original experiment of Fig. 1, irons of varying degree of abnormality were poured in the open molds of the three sands described. No marked effect of the mold on graphite distribution or the character of the matrix could be detected in the 1.2-in. diameter round section in either the normal or abnormal irons. Small wedge chill specimens, poured in the three sands, showed less chill in the ignited molds, but this effect could not be found in the 1.2-in. diameter round section. The amount of pipe in the bars, or the effect of sand on the piping tendency, was not appreciably influenced by changing the iron from normal to abnormal by holding in the furnace and superheating. It should be noted that these conclusions regarding the effect of sand on the microstructure of the 1.2-in. diameter section apply only to castings made in open molds.

Theory Advanced

58. Carmody[1], has called attention to the elimination of

[1] Carmody, E. J., *"Preventing Shrinks and Blowholes,"* THE FOUNDRY, vol. 66, no. 11, November 1938, pp. 32, 33, 84, 87.

shrinkage defects by expansion of cores in the molds. The apparent absence of any appreciable temperature effect or influence on microstructure, suggests that the differences in the case of the various green sands can be explained as a result of expansion or contraction of mold material. During solidification, gray iron passes through a highly plastic state. It would appear that the solid metal in a partially solidified casting behaves as a plastic membrane which follows any movement of the mold wall occurring as a result of the action of high temperature and ferrostatic pressure on the sand. Resulting volume changes are fed by molten metal from the riser. In the absence of liquid iron in a riser, shrink cavities are formed, or sinking of the cope face of the casting occurs.

59. Attempts have been made to measure the movement of mold walls under ferrostatic pressure using an indicating device attached to the mold-metal interface. A number of curves have been obtained which show an expansion of the core sand and a contraction of the green sand occurring after pouring has ceased. These results cannot be reproduced consistently and are not reported.

Change in Dimension and Weight of Test Pieces

60. Table 1 gives data obtained by measuring several groups of the heavy section castings on the dimension marked 4½-in. in Fig. 8. These figures show a correlation between shrinkage and casting dimension, and tend to substantiate the theory that piping and shrinkage are associated with movement of the mold wall during solidification.

Table 1

CHANGE OF 4½-IN. DIMENSION OF FIG. 8 WITH VARIOUS DEGREES OF PIPING

Figure No.	*Dimension with Small Pipe, in.*	*Dimension with Medium Pipe, in.*	*Dimension with Large Pipe, in.*
10	4.47	4.52	4.53
12	4.45	4.47	4.52
15	4.44	4.45	4.45
19	4.45	4.48	4.53

A similar correlation has been found between the weight of the castings and the amount of piping. Deeply piped castings are

generally slightly heavier than castings from the same ladle showing less pipe.

CONCLUSIONS

61. During this investigation, progress reports have been submitted to members of the Foundry Sand Research Committee of the Association. This material is presented as a progress report at the suggestion of the committee.

62. The following indications are given:

1. Differences in piping, internal shrinks, and sinking of the cope face of gray iron castings are correlated with dimensional changes of the mold cavity, resulting from expansion and contraction of the mold material during solidification of the metal.

2. High moisture, high green compressive strength, soft ramming, and unfavorable distribution of grain size of the sand, tend to promote shrinkage defects in gray iron castings.

3. The presence of sea coal in green sand decreases the amount of piping in risers in gray iron castings.

4. Small gray iron castings made in baked oil sand molds show no shrinkage defects.

5. Small steel castings are less susceptible to the influence of the mold than are gray iron castings.

63. It should be emphasized that the statement regarding the influence of green compressive strength on shrinkage applies only to the conditions prevailing in the experiments as outlined in this report.

ACKNOWLEDGMENTS

64. The initial experiment of this investigation was performed as a senior project in April 1938, by John Sangster and Orlo Carlson, students at Michigan State College. The research has been discussed before a committee representing the Detroit chapter of A. F. A., and was delivered as a progress report before a regional meeting of the Association at Michigan State College, April 1939.

DISCUSSION

Presiding: W. G. REICHERT, American Brake Shoe & Foundry Co., Mahwah, N J.

H. W. DIETERT[1]: Do the authors feel that some of the piping might have been due to deformation or swelling of the mold? For example, when bentonite is added to the molding sand and then the moisture is quite high, you have high deformation. That might lead one to assume that the swelling or the deformation of the mold enlarged the cavity in which the iron solidified thus causing the piping. Did that deformation have something to do with this phenomena?

MR. WOMOCHEL: We feel that is the case.

F. G. SEFING:[2] It occurs to me that a good deal of the shrinkage that was evidenced in their results might be due to an increase in the density of the metal itself, and I would suggest that the castings which they still have be run for specific gravity to correlate with their shrinkage phenomena. I would suggest too, that they test tensile strengths on the irons that show different shrinkage properties. It has been my experience that when irons pull down in the riser nicely, one need not worry about draws and shrinks. If the metal stays right up in the top of the riser, we are considerably worried about the casting. I do not know exactly what causes these differences but I wonder whether the properties of the metal are not superior when we get proper shrinkage. I wonder too, whether the density of the metal is not improved with proper shrinkage. Therefore, I am suggesting this additional test work to complement the information of this paper.

MR. WOMOCHEL: During the progress of this work, we considered all the possibilities we could think of, and we finally arrived at the conclusion that most of the shrinkage was due to the expansion and contraction of the mold. No doubt there are many factors operating to affect shrinkage. We have made a limited number of density determinations on sections of castings taken from deep pipe and shallow pipe molds, and at the present time we have not been able to correlate any changes in the density of the metal with the difference in piping tendency observed. However, some of the castings showed porous areas at the base of the risers, and no doubt the metal at those points is less dense than the metal taken from other parts of the casting. We regarded this porous condition at the base of the riser as a less serious manifestation of the condition that caused an actual cavity.

H. S. WASHBURN:[3] Have the authors formulated any idea in regard to this subject of minute particles of gas in the iron?

MR. WOMOCHEL: We have considered that possibility, and early during the research, it appeared to us that gas in the metal was partly

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[1]Harry W. Dietert Co., Detroit, Mich.
[2]International Nickel Co., New York, N. Y.
[3]President, Plainville Casting Co., Plainville, Conn.

responsible for the conditions that we observed, but we have not been able to confirm that at the present time. Our purpose in making the density determinations was to determine whether there were some microscopic gas inclusions in the metal.

N J. DUNBECK:[4] Mr. Dietert's statement about deformation in this sort of work is important. It is more important on an actual practical mix for use in your foundry than in what we would call a theoretical mixture. If we have a sand that has bad expansion or contraction conditions, we must correct that in the foundry to make a good casting. Therefore, we can assume that the average sand in the average foundry has fairly good expansion and contraction properties. On the other hand, two sands, which show very good properties in all the other tests we make, can show a difference of as much as 300 per cent in hot deformation. I suggest hot deformation may be one of the important reasons for changes in the volume of the size of this mold. I understand the authors are going to do more work on that. I think it should be very helpful and informative.

E. J. ASH[5] *(Written Discussion):* It is apparent that the authors have expended considerable effort in making the tests outlined in this paper. It is interesting to analyze the results obtained with regards to explaining the cause of the apparent shrinkage obtained in both cast iron and steel.

It is obvious that there are two primary factors in these tests that can influence the apparent amount of voids created in the respective castings, namely, (1) rate of cooling and (2) dimensional change of the mold cavity during casting.

It is probable that different sand mixtures may confer a difference in the cooling rate of the cast iron which, in turn, will effect the volume change during solidification (slower cooling would tend to permit more complete breakdown of cementite). A high permeability or high moisture in the sand will effect this variable. In these tests, however, it is doubtful whether these variables had any measurable influence in controlling the state of the carbon.

The second factor influencing the total amount of apparent voids exhibited by the cast iron, and undoubtedly the major factor in these tests, is influenced by a number of variables. Deformation, flowability, expansion and contraction characteristics of the sand, together with ferrostatic pressure and solidification thrust, are the dominating variables that would tend to produce an increase in mold dimension during casting.

The following examples will serve to indicate the influence of the above variables:

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[4] Vice president, Eastern Clay Products Co., Eifort, O.

[5] Assistant Professor, Dept. of Metal Processing, University of Michigan, Ann Arbor, Mich.

In Fig. 4, the highest strength sand produced the largest cavity because of the flowability change in the sand. Flowability of a molding sand decreases with increase in clay content (strength) hence, the mold high in strength was not rammed as firmly as the molds of the lower strengths, resulting in increase in swell as the strength of sand increases. This is further illustrated by the dimensions given in Table 1, paragraph 60. The casting with the larger pipe was larger in size than the other respective castings. Likewise, in Fig. 5, the higher the permeability, the lower the flowability, especially when dealing with values under 100. In Fig. 6, moisture also effects flowability, the higher the moisture content (to temper), the lower is the flowability.

It is not inferred that change in flowability, ferrostatic pressure and solidification thrust are the only variables contributing to the mold cavity changes, but there may be other variables. In Fig. 7, the addition of seacoal has a very marked effect in eliminating the voids created in the cast specimen. Seacoal additions decrease the flowability of a molding sand, but they also decrease the expansion and contraction of the sand. The latter variables were far more dominating than the effect of flowability, resulting in a casting that did not change dimensions appreciably during pouring.

Core sand molds yielded solid castings because of the rigidity of the cores, their ability to withstand the ferrostatic pressure when the mold is first poured, and also to withstand the expanding thrust associated with the expansion of the metal during solidification.

That the change in size of the mold cavity is a factor can be deduced from the change in dimensions given in Table 1. The data for Fig. 10 are given:

Dimension With Small Pipe, In.	*Area, Sq. In.*	*Dimension With Large Pipe, In.*	*Area, Sq. In.*	*Difference Sq. In.*
4.47	15.693	4.53	16.117	0.424

When the above difference in cross-sectional area (0.424) is multiplied by the third dimension (diameter) of the specimen, a considerable volume (void) is shown to be created.

Steel results appear not to be affected. This is partly due to the rapid rate at which a rigid steel shell forms, since the temperature of solidification is higher, temperature differential between mold and metal is greater, resulting in a greater heat dissipation. Furthermore, on freezing, steel does not offer the solidification thrust associated with the freezing of gray cast iron. Solid contraction curves of the two respective materials demonstrate this.

It should be emphasized that the molding material itself has no effect whatever on the shrinkage characteristics of a metal. Any such observed differences are apparent and are associated with either the rate of cooling or mold cavity change as pointed out previously.

It, therefore, appears that there are several additional measurable variables not determined in these tests that would have an important

bearing on the apparent voids obtained. Measurement of flowability, deformation, expansion, and contraction of the sands would have, undoubtedly, aided in explaining the difference in voids created by gray cast iron during freezing.

H. WOMOCHEL and C. C. SIGERFOOS *(Written Closure):* Professor Ash's statements, regarding the effect of dimensional changes of the mold cavity on apparent shrinkage, are in exact agreement with our views as presented in the preprint. While engaged in this work we considered all the obvious factors which might contribute to changes in the volume of the cavity. Among these was flowability of the sand. Other causes, which might be mentioned, are contraction of the sand, resulting from evaporation of moisture, and from contraction of the clay, during its dehydration at the high temperatures existing in the mold. In view of the many possible explanations which can be advanced, we do not feel that Professor Ash is justified in attributing the results obtained primarily to lack of flowability of the sands used.

The patterns, employed in this work, were round or spherical shape to minimize soft spots due to lack of flowability. The molding procedure used was a combination of hand ramming and squeezing in shallow flask sections to further reduce lack of uniformity of hardness to a minimum. Test molds were sectioned, as described in paragraph 24, to assure ourselves that lack of flowability was not causing soft spots in the molds.

It should not be assumed that the factors mentioned by Professor Ash were not considered by the authors. Much speculation of this nature was intentionally omitted from this progress report. Our attempt has been to demonstrate that such factors as moisture content, green compressive strength, and sea coal content have an influence on the size and distribution of casting defects usually attributed solely to volume change in the metal. That variations in these factors are accompanied by changes in deformation, hot contraction, and flowability is of course obvious to all foundrymen. Moisture, strength, permeability and mold hardness were selected as the basis for our experiments in preference to flowability, deformation, expansion, and contraction because the former quantities are those ordinarily measured in the foundry.

Desulphurizing Cupola Cast Iron From the Practical Operating Angle

By W. Levi*, Lynchburg, Va.

1. The purpose of this paper is to discuss the actual practice followed in the desulphurizing of cupola cast iron in the foundries of the company with which the author is connected.

2. The castings manufactured include a complete line of plow castings such as chilled plow points or shares, chilled landsides, standards, and other parts for plows. In addition, cast iron pipe ranging in size from 3-in. to over 60-in. in diameter is made by the pit cast method or in loam, deLavaud centrifugally cast pipe from 4 through 24-in. in diameter, practically all sizes of fittings for gas, water and steam service, and a wide variety of special castings for use in the chemical industry.

3. Historically, desulphurizing cast iron is not a particularly new procedure. In fact, we are told that Horsley discussed the use in England of soda ash and other chemicals for this purpose as long ago as 1867. His chief interest at that time, however, was the effect of soda ash refinement upon the grain structure and density of the iron treated rather than the reduction of sulphur.

Melting Equipment

4. The melting equipment used for production castings consists of six cupolas all equipped with continuous front slagging spouts and either a forehearth or some type of teapot ladle used for the storage and mixing of iron and refining with fused soda ash. The cupolas referred to are as follows:

* Metallurgist, Lynchburg Foundry Company.

Note: This paper was presented at a Gray Iron Session of the A.F.A. Convention, May 9, Chicago, Ill.

No. of Cupolas	*I.D. Shell-In.*	*I.D. Lining at Melting Zone, In.*	*Forehearth*	
			Type	*Capacity*
1	63	51	Box	3200 lb.
1	54	42	U-shaped Teapot	2 tons
1	96	72	U-shaped Teapot	10 tons
2	96	72	Elliptical Teapot	10 tons
1	96	66	U-shaped Teapot	10 tons

5. Due to the geographic location of the foundries in question, the use of large percentages of scrap iron in the cupola charge is of vital importance in order to maintain an iron cost which is not prohibitive. The amount of scrap iron used in the various mixtures ranges from about 83 per cent to 100 per cent. With proper supervision and control, irons containing from 2.90 to 3.75 per cent total carbon have been made repeatedly using the cupola mixtures mentioned.

6. In discussing the refining and desulphurizing of cupola cast iron by means of fused soda ash, it must be borne in mind that every detail of good cupola practice must be given considerable thought and consideration as usual.

Preheating Forehearth

7. The procedure followed in our plants for melting a 100 per cent scrap charge consists in preheating the forehearth or teapot ladle with an oil torch for 1½ to 3 hr. before the blast is turned on the cupola. This time varies according to the size of the forehearth and the type of burner used. However, in all cases it is important to have the forehearth lining at a minimum temperature of 2400°F. before iron is tapped into it. A superficial heating of the forehearth lining is often the cause of dull iron at the beginning of a heat and for this reason this step in the preparation cannot be over-estimated. Further, the burner used for preheating should not be removed until just before tapping the cupola. The purpose of this is, of course, to conserve the heat already absorbed by the brick in the forehearth lining.

Addition of Desulphurizer

8. After the iron in the forehearth has attained a depth sufficient to seal or cover the hole leading from the forehearth proper to the teapot spout, a predetermined amount of fused soda ash is added. This amount is governed by previous experience and also by the capacity of the forehearth, the object being to add a definite predetermined amount of fused soda ash per ton of molten iron.

9. At this point, it might be well to state that before any fused soda ash is added to the molten iron, it should be dried sufficiently to drive out any absorbed moisture. This can be accomplished by placing a day's supply in a suitable oven, by placing the pieces over the cupola spout and replacing dried ones with undried ones as they are used, or by means of the flame of an oil or gas torch played on the surface of a stock pile for a few minutes prior to use.

10. When the forehearth is full, the iron is either tapped out or poured into a thoroughly-dried and well-preheated transfer ladle and is then taken to the pouring floor. Immediately after pouring iron into the transfer ladle, sufficient fused soda ash is added to the forehearth to treat an amount of iron equal to that which was removed. This is very important as any delay at this point reduces the time that the iron running into the forehearth is in contact with the purifying slag. When conditions are such that the ratio of the capacity of the forehearth to the cupola melting rate is small, any delay will mean that a relatively large percentage of the available contact time has been lost and consequently the efficiency of the desulphurizing equipment reduced. Succeeding ladles of iron should not be taken out until the forehearth has been allowed to completely fill again. After several ladles have been taken out and a like number of additions of fused soda ash have been made to the forehearth, a layer of slag will have formed on the surface of the metal. This used slag should be skimmed off by allowing the level of the iron in the forehearth to rise to a point just below the slag spout. It has been noted that more efficient desulphurizing is obtained when there is a thin layer of slag on the surface of the iron.

11. This cycle of adding fused soda ash to the forehearth, allowing the forehearth to become completely filled, skimming off the slag, and then pouring iron into the transfer ladle is continued throughout the duration of the day's cast. However, it is not necessary to add soda ash for the last ladle or two as the layer of refining slag already present will continue to desulphurize for at least 10 to 15 min.

Factors Influencing Desulphurization

12. The amount of sulphur that will be present in a cupola-melted iron after treatment with fused soda ash is a function of

a good many factors some of which are listed below:

1. Type of desulphurizing reservoir used.

2. "Contact time" between desulphurizing agent and molten iron. "Contact time" in minutes is defined as the ratio of the forehearth capacity to the hourly melting rate of the cupola multiplied by sixty (60).

3. Amount of sulphur present in the iron at the cupola spout (later referred to as "base metal") before desulphurizing. This, in turn, is proportional to the amount of sulphur present in both the coke and metals charged into the cupola.

4. Cupola melting rate.

5. Coke ratio used for melting.

6. Amount of desulphurizing agent used per ton of molten iron.

7. Completeness of the separation of the cupola slag from the molten iron before desulphurizing.

Effect of Type Reservoir on Sulphur Removal

13. To demonstrate the effect on sulphur removal of two different types of desulphurizing reservoirs as well as some difference in contact time, let us consider the data given in Fig. 1 and Fig. 2. While the cupola charges and coke ratios are not identical in the two units being considered, the sulphur averages in the irons at the spouts are practically identical.

CASE NO. 1

Cupola Dimensions

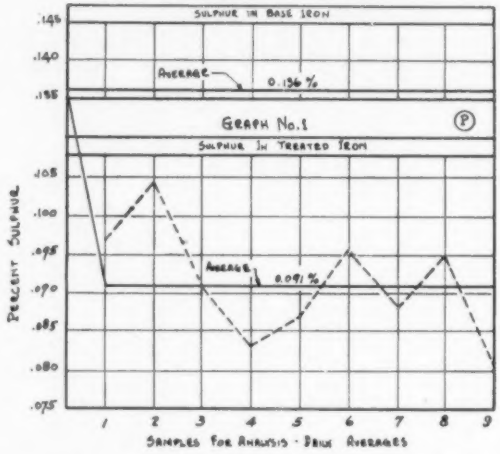
14. The cupola discussed in Fig. 1 is used to melt iron for chilled plow points (or shares), chilled landside castings, standards, and other plow castings. The inside diameter of the cupola is 51 in. at the melting zone, and the average melting rate is 7.5 tons per hr. The cupola charge contains 12.9 per cent pig iron.

Desulphurizing Reservoir

15. The "box" type reservoir used in connection with this cupola is rectangular in cross-section, the dimensions inside the lining being 32 in. long, 24 in. wide, and 18 in. deep, with a capac-

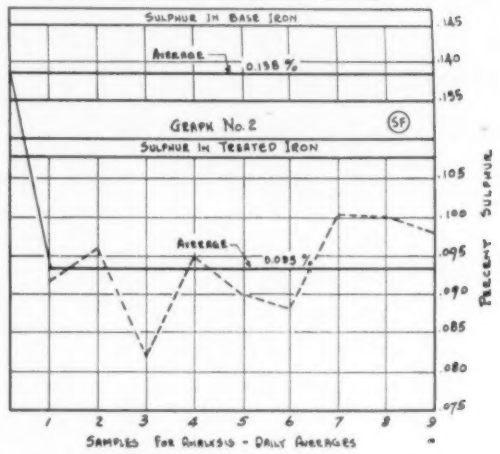
DESULPHURISING CUPOLA CAST IRON

BOX TYPE - 3200 LB FOREHEARTH



FUSED SODA ASH PER TON	11.6 LBS.	IRON TEMPS.	°F
REDUCTION IN S.	.045 %	CUPOLA SPOUT	2831-2846
SULPHUR REMOVED	33 %	BOX SPOUT	2702-2718
S PER LB. FUSED SODA ASH	.00388 %	TEMP. LOSS	128
COKE RATIO	6.2 TO 1	CUPOLA MIXTURE	
MELTING RATE	7.5 TONS	PIG IRON	13.0 %
CONTACT TIME	12.8 MINS.	CAST SCRAP	80.7 %
		STEEL SCRAP	6.4 %

ELONGATED U-SHAPED 2 TON T-POT LADLE



FUSED SODA ASH PER TON	8 LBS.	IRON TEMPS.	°F
REDUCTION IN S.	.045 %	CUPOLA SPOUT	2702-2766
SULPHUR REMOVED	32.6 %	T-POT SPOUT	2559-2684
S PER LB FUSED SODA ASH	.0056 %	TEMP LOSS	82-143
COKE RATIO	6.0 TO 1	CUPOLA MIXTURE	
MELTING RATE	7 TONS	PIG IRON	16.7 %
CONTACT TIME	17 MIN.	CAST SCRAP	77.1 %
		STEEL SCRAP	6.2 %

FIG. 1—LEFT—RESULTS OBTAINED BY DESULPHURIZATION IN "BOX" TYPE DESULPHURIZING LADLE. FIG. 2—RIGHT—RESULTS OBTAINED IN 2-TON CAPACITY, ELONGATED, U-SHAPED, TEAPOT-TYPE TILTING LADLE.

ity of 3200 lb. of molten iron. It has a brick lining 7-in. thick and the cover is lined with 4-in. of ganister. Fused soda ash additions are made through the same opening that the spent purifying slag is skimmed off from the surface of the iron in the forehearth, that is, through the slag spout. This forehearth is of the stationary type and must be "tapped out" each time iron is taken from it.

16. The analysis of the iron at the cupola spout is as follows:

Element	*Per Cent*
Silicon	1.40 –1.60
Sulphur	0.125–0.155
Manganese	0.40 –0.50
Phosphorus	0.45 –0.60
Total Carbon	3.30 –3.50

Sulphur Reduction

17. The average sulphur in the base metal is 0.136 per cent and, after treatment with 11.6 lb. of fused soda ash per ton, the sulphur is reduced to an average of 0.091 per cent. The available contact time for the refining action of the fused soda ash is 12.8 min.

Costs

18. The cost of soda ash per ton of good plow castings produced is $0.399. The cost of refractories, for the forehearth lining, which includes labor and materials, is $0.039 per ton of iron melted. This refractory cost is an average of the performance of seven linings. The average figures are as follows:

Tons	*Refractory Costs*			
Melted	*Labor*	*Materials*	*Total*	*Per Ton*
428	$8.50	$8.23	$16.73	$0.039

Coke Analysis

19. The coke used in this cupola is of the by-product type and is of the following analysis:

	Per Cent
Fixed Carbon	92.0
Sulphur	0.60
Ash	6.0
Volatile	0.65

CASE NO. 2

20. For the data in Fig. 2, the cupola is used to melt iron for

the production of short length flanged pipe, bell and spigot pipe, fittings, and special castings. The maximum length of pipe made rarely exceeds 5 ft. overall and the maximum diameter of pipe and fittings manufactured is 20 in.

Cupola Dimensions

21. The inside diameter of the cupola is 42 in. and the average melting rate is 7 tons per hr. The cupola charge contains 16.7 per cent pig iron.

Desulphurizing Reservoir

22. The desulphurizing reservoir used is a 2-ton capacity, elongated, U-shaped, teapot type tilting ladle. It is lined with 9-in. of firebrick, and the cover is lined with 6½-in. of ganister. The depth of molten iron in the teapot ladle when full is 21 in.

Analysis of Iron

23. The analysis of the iron at the cupola spout is as follows:

Element	*Per Cent*
Silicon	1.65 –1.85
Sulphur	0.125–0.160
Manganese	0.35 –0.50
Phosphorus	0.45 –0.60
Total Carbon	3.30 –3.50

Sulphur Reduction

24. The average sulphur in the base metal is 0.138 per cent and after treatment with 8 lb. of fused soda ash per ton, it is reduced to an average of 0.093 per cent. The available contact time for the refining action of the fused soda ash is 17 min.

25. A comparison of results shown in Figs. 1 and 2 indicates that the sulphur reduction with the "box" type forehearth is 0.00388 per cent per lb. of soda ash, and with the elongated U-shaped teapot ladle the sulphur reduction is 0.0056 per cent per lb. of soda ash. This difference in sulphur reduction is due both to the type of desulphurizing reservoir and to the available contact time between the soda ash and the molten iron. Sulphur reduction is proportional to contact time and, when desulphurizing equipment is being installed, this phase should be carefully considered. The greater efficiency of the elongated type forehearth is due to the fact that the boiling action of the soda ash is limited to the end where the iron enters, and that the refined

metal is tapped out at the opposite end. While the iron travels from one end of the ladle to the other, the soda ash slag rises to the surface and removes sulphur and entrained impurities.

26. The cost of the desulphurizing material per ton of good castings using the 2-ton, elongated, teapot ladle as the refining reservoir is $0.282. The cost of refractories used in the forehearth, including both labor and materials, is $0.046 per ton of iron melted. This cost is an average of the performance of six linings, the average figures being as follows:

Tons	*Refractory Costs*			
Melted	*Labor*	*Materials*	*Total*	*Per Ton*
726	$11.57	$21.78	$33.35	$0.046

27. By-product coke of the same analysis shown in paragraph 19 is used in this cupola.

COMPARISON OF 10-TON TEAPOT LADLES

28. The data given in Figs. 3 and 4 are intended to point out the relative desulphurizing efficiencies of two different types of 10-ton, teapot, tilting ladles. All other conditions are as near identical as possible with an operation of this size.

29. In both cases, the iron melted is used for the production of pit-cast pipe, fittings over 20-in. in diameter, and a large variety of special shapes for use in the chemical industry.

Cupola Dimensions

30. Both cupola shells under consideration are 96-in. inside diameter and are lined to 72-in. at the melting zone, and the melting rate is 16-tons per hr. The 4000 lb. charge consists of 100 per cent scrap as follows:

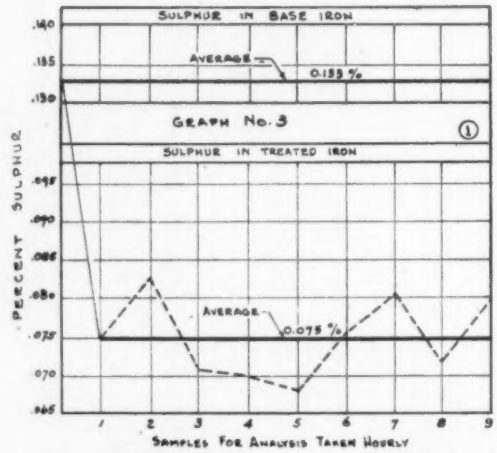
Type of Scrap	*Lb.*	*Per Cent*
Heavy Machinery	1200	30.0
Automotive Cast	900	22.5
Returns	700	17.5
Cotton Mill	400	10.0
No. 1 Machinery	800	20.0
Totals	4000	100.0

Coke Analysis

31. Coke between charges consists of 450 lb. of beehive coke and 25 lb. of pitch coke. The analyses of the cokes used are as follows:

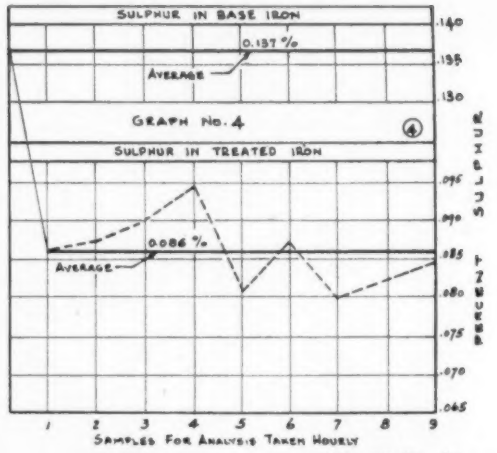
DESULPHURISING CUPOLA CAST IRON - 100 % SCRAP CHARGE

ELONGATED U-SHAPED 10 TON T-POT LADLE



DATA		IRON TEMPS. °F.	
FUSED SODA ASH PER TON	5.8 LBS.	CUPOLA SPOUT	2676 - 2722
REDUCTION IN S.	.068 %	T-POT SPOUT	2569 - 2626
SULPHUR REMOVED	48.5 %	TEMP. LOSS	96 - 107
S PER LB. FUSED SODA ASH	.010 %		
COKE RATIO	8.4 TO 1		
MELTING RATE	16 TONS		
CONTACT TIME	37.5 MIN.		

STANDARD ELLIPTICAL SHAPED 10 TON T-POT LADLE



DATA		IRON TEMPS. °F.	
FUSED SODA ASH PER TON	6.4 LBS.	CUPOLA SPOUT	2687 - 2705
REDUCTION IN S.	.051 %	T-POT SPOUT	2589 - 2607
SULPHUR REMOVED	37.2 %	TEMP. LOSS	98
S PER LB FUSED SODA ASH	.0079 %		
COKE RATIO	8.4 TO 1		
MELTING RATE	16 TONS		
CONTACT TIME	37.5 MIN.		

RELATIVE DESULPHURIZING EFFICIENCIES OF TWO, 10-TON, TEAPOT, TILTING LADLES. FIG. 3—LEFT—RESULTS OBTAINED ON ELONGATED, U-SHAPED, TEAPOT, TILTING LADLE. FIG. 4—RIGHT—RESULTS OBTAINED ON ELLIPTICAL-SHAPED, TEAPOT-TYPE, TILTING LADLE.

	Kind of Coke	
	Beehive	*Pitch*
Fixed Carbon, per cent	93.5	98.0
Sulphur, per cent	0.46	0.30
Ash, per cent	5.0	0.50
Volatile, per cent	0.65	1.00

32. The average cost of fused soda ash per ton of good castings produced using either the U-shaped or the elliptical shaped forehearth varies from $0.20 to $0.26. The U-shaped ladle requires 0.6 lb. less of the fused soda ash per ton than does the elliptical ladle, and at the same time removes more sulphur. The cost also varies with the type of castings being produced. Where gates and runners constitute a large percentage of the iron poured into a mold, the cost of the desulphurizing material will be higher than when this percentage is small. In other words, the cost per ton of good castings will be inversely proportional to the ratio of good yield to total iron melted.

CASE NO. 3

33. The desulphurizing reservoir used for the data in Fig. 3 is a 10-ton capacity, elongated, U-shaped, teapot type of tilting ladle. It has a firebrick lining 11½-in. thick and the cover is lined with 6½-in. of ganister. The depth of molten iron in the teapot ladle when full is 36 in.

34. Fused soda ash additions are made to the teapot ladle at the same end that the molten iron enters and the spent slag is skimmed off at the opposite end. The refined iron is also tapped out at the end opposite that at which the molten iron enters.

Analysis of Iron

35. The analysis of the iron at the cupola spout is as follows:

Element	*Per Cent*
Silicon	1.50 –1.70
Sulphur	0.130–0.150
Phosphorus	0.45 –0.60
Manganese	0.35 –0.45
Total Carbon	3.30 –3.60

Sulphur Reduction

36. The average sulphur in the base metal is 0.133 per cent and after treatment with 5.8 lb. of fused soda ash per ton, the

sulphur is reduced to an average of 0.075 per cent. The available contact time for the refining action of the fused soda ash is 37½ min.

Costs

37. The cost of refractories for the 10-ton, elongated, U-shaped, teapot ladle lining, including labor and materials, is $0.041 per ton of iron melted. This cost is an average of the performance of nine linings and the average figures are as follows:

Tons	*Refractory Costs*			
Melted	*Labor*	*Materials*	*Total*	*Per Ton*
1736	$15.76	$56.26	$72.02	$0.041

CASE NO. 4

38. As stated herein (paragraphs 28 to 37 inclusive) the only difference in conditions relative to data covered by Fig. 3 and Fig. 4 is the type of desulphurizing reservoir used. Here a 10-ton capacity, elliptical-shaped, teapot type of tilting ladle is used. It has a 7¾-in. firebrick lining and no cover is used. The depth of molten iron in the teapot ladle when full is about 27 in.

39. Fused soda ash additions are made at or near the center of the teapot ladle. This is close to the slag spout and also quite close to the point at which molten iron is removed from the ladle.

Sulphur Removal

40. The average sulphur in the base metal is 0.137 per cent and, after treatment with 6.4 lb. of fused soda ash per ton, the sulphur is reduced to 0.086 per cent. The available contact time for the refining action of the fused soda ash is 37½ min.

41. The refractory cost, including labor and materials for lining the 10-ton, elliptical-shaped, teapot is $0.43 per ton of iron melted. The figure given is an average of the performance of ten linings, and the average figures are as follows:

Tons	*Refractory Costs*			
Melted	*Labor*	*Materials*	*Total*	*Per Ton*
1209	$12.96	$39.03	$51.99	$0.043

42. The sulphur reduction, when using the elongated type teapot ladle is 0.010 per cent per lb. of fused soda ash and, when using the elliptical type ladle, the reduction is only 0.0079 per cent. This difference is accounted for solely by the difference in types of desulphurizing reservoirs used.

EFFECT OF COKE RATIO ON SULPHUR CONTENT

43. The purpose of Figs. 5 and 6 is to point out the effect of coke ratio on the sulphur content of the iron at the cupola spout, and, in turn, its effect on the amount of desulphurizing material used.

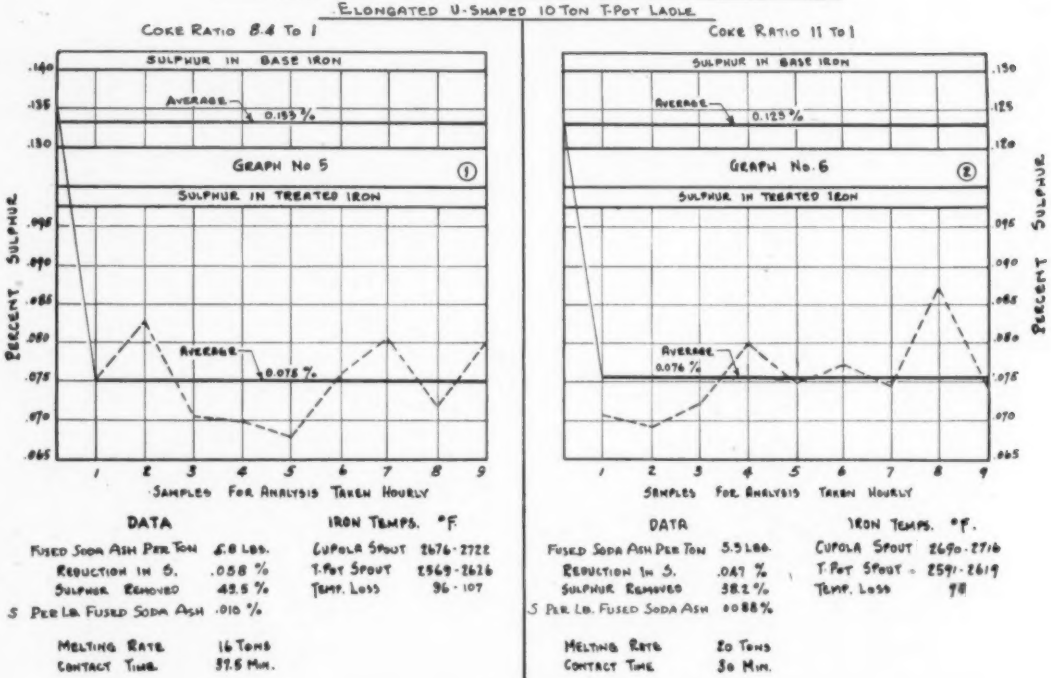
44. In both cases, desulphurizing is done in the same 10-ton, elongated, U-shaped, teapot ladle that was discussed under Fig. 3 (paragraphs 28 to 37 inclusive). Fig. 5 is the same as Fig. 3 and naturally all data pertaining to Fig. 3 applies to Fig. 5.

45. To obtain the data presented in Fig. 6, the coke ratio was changed to 11 to 1 as compared to a coke ratio of 8.4 to 1 in Fig. 5. This is the only variable introduced deliberately. However, this has resulted in increasing the melting rate from 16 tons per hour to 20 tons per hour, and this in turn has reduced the time of contact between the molten iron and the fused soda ash slag from 37.5 min. to 30 min. It has also reduced the average sulphur in the iron on the cupola spout (base metal) from 0.133 to 0.123 per cent.

46. Examination of the data presented in Fig. 5 shows the sulphur reduction per pound of fused soda ash to be 0.010 per cent. Fig. 6 shows the sulphur in the base metal to be 0.010 per cent lower than the base metal sulphur in Fig. 5, and on this basis, one might assume that, to arrive at the same sulphur content in the desulphurized iron, it would be satisfactory to use one pound less of fused soda ash per ton under the conditions described in Fig. 6 than under those described in Fig. 5. However, from actual practice we find that to arrive at the same sulphur content in both treated irons, we are able to make a reduction in the amount of fused soda ash used of only ½-lb. per ton. (The actual reduction is from 5.8 lb. to 5.3 lb.) This is accounted for by the reduction in contact time resulting from the increased melting rate.

DESULPHURIZING VERSUS FLUIDITY

47. The effect of desulphurizing cupola cast iron on its fluidity is a subject of considerable controversy and it is the belief of some that the sulphur reduction in no way effects this property. It does appear, however, that the removal of entrained



EFFECT OF COKE RATIO ON SULPHUR CONTENT OF IRON AT THE SPOUT. FIG. 5—LEFT—SULPHUR CONTENT WITH COKE RATIO OF 8.4 TO 1. FIG. 6—RIGHT SULPHUR CONTENT WITH COKE RATIO OF 11 TO 1.

silicates and cupola slag coming out of the cupola with the molten iron does increase the fluidity. Removal of these undesirable materials from the molten iron takes place simultaneously with desulphurizing and we are told that it requires about 1-lb. of fused soda ash per ton of molten iron to accomplish this.

48. In making fluidity tests, the standard A.F.A. fluidity test pattern was used. All molds were rammed from the same sand heap which had a permeability of 38, a green shear of 1.2-lb. and a moisture content of 6.9 per cent. A sample of desulphurized iron was caught in a 50-lb. hand ladle and poured into one of the molds at a temperature of 2400°F. Next a sample of iron was caught from the cupola spout and poured at 2400°F. into a second fluidity test mold. It is obvious, from the procedure just described, that the sulphur content of the base metal will be higher than that of the desulphurized iron. In addition, the individual determinations of silicon, manganese, phosphorus, and total carbon made on the base iron are not identical with those made on the corresponding sample of treated iron. However, the average analyses of the base iron and the desulphurized iron are, with the exception of sulphur content, very nearly alike.

49. Data covering fluidity tests on cupola cast iron are shown in Fig. 7. These data were taken at the same time some of the data shown in Fig. 4 were obtained and consequently the same conditions apply in both cases. The sulphur reduction obtained was from an average of 0.137 per cent to an average of 0.086 per cent with the use of 6.4 lb. of fused soda ash per ton of molten iron. The increase in fluidity was from 27.3 in. (average) to 36.4 in. (average) in the desulphurized iron. This means an average increase in length of the fluidity test strip of 9.1 in. or 33.3 per cent.

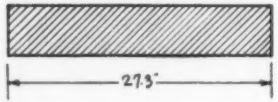
50. It is not to be understood that to obtain an increase in fluidity of 33.3 per cent requires the use of 6.4 lb. of fused soda ash per ton of metal, as increases of the same order of magnitude have been obtained with the use of 4-lb. per ton.

— DESULPHURISING CUPOLA CAST IRON - 100% SCRAP CHARGE —

EFFECT ON FLUIDITY - ALL TEST STRIPS POURED AT 2400°F.

FLUIDITY OF BASE IRON

AVERAGE LENGTH OF FLUIDITY TEST STRIP - 27.3"

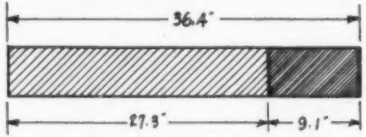


ANALYSES OF IRONS USED FOR FLUIDITY TESTS

TEST	% Si.	% S.	% P.	% Mn.	% T.C.	FLUIDITY
1	1.62	.137	.48	.32	3.57	26"
2	1.57	.135	.62	.48	3.46	25"
3	1.58	.133	.69	.51	3.40	26"
4	1.44	.145	.49	.41	3.60	24"
5	1.61	.141	.40	.47	3.64	28"
6	1.59	.137	.60	.34	3.49	29"
7	1.63	.132	.46	.39	3.66	27"
8	1.72	.145	.48	.37	3.44	30"
9	1.67	.130	.51	.40	3.58	31"
Av.	1.60	.137	.52	.41	3.54	27.3"

FLUIDITY OF IRON TREATED WITH 6.4 LBS. FUSED SODA ASH PER TON

AVERAGE LENGTH OF FLUIDITY TEST STRIP - 36.4"



INCREASE IN FLUIDITY IS 9.1" OR 33.3%

ANALYSES OF IRONS USED FOR FLUIDITY TESTS

TEST	% Si.	% S.	% P.	% Mn.	% T.C.	FLUIDITY
1	1.53	.086	.63	.38	3.52	36"
2	1.59	.087	.57	.41	3.51	34"
3	1.52	.090	.60	.38	3.49	32"
4	1.54	.094	.53	.36	3.47	35"
5	1.56	.081	.57	.38	3.50	38"
6	1.66	.087	.52	.40	3.55	37"
7	1.58	.080	.50	.38	3.60	40"
8	1.57	.082	.52	.40	3.52	38"
9	1.55	.084	.48	.40	3.54	38"
Av.	1.56	.086	.54	.39	3.51	36.4"

FIG. 7—EFFECT OF ADDITION OF FUSED SODA ASH TO CUPOLA IRON ON FLUIDITY.

DISCUSSION

In absence of the author the paper was presented by M. Kuniansky, Lynchburg Foundry Co., Lynchburg, Va.

Presiding[1]: G. P. PHILLIPS, Automotive Foundry Div., International Harvester Co., Chicago, Ill.

H. BORNSTEIN:[1] I would like to ask Mr. Kuniansky why so much soda ash is used. Is there anything particularly gained in going down as low in sulphur as they do? Why is not 0.10 or 0.11 per cent sulphur satisfactory? What other benefits are derived from large additions of fused soda ash?

MR. KUNIANSKY: A lot of our specifications call for maximum sulphur of 0.10 per cent. Quite obviously we do not try to just hit below the mark; we try to give plenty of leeway.

In the manufacture of super-deLavaud pipe, which is a product that has to be annealed, we try to run a very low manganese, preferable 0.40 per cent or under, and inasmuch as we try to keep a 5 to 1 ratio of manganese to sulphur it, therefore, becomes necessary to keep sulphur around 0.075 per cent or under.

We have been following this practice for about nine years in the course of which we have made about 400,000 tons of castings, and we believe that we get improved machinability in our machine shops and have fewer rejections due to the more extensive treatment that we give it. We have not any too much margin.

CHAIRMAN PHILLIPS: One point the author brought out is that the U-shaped ladle is certainly much more efficient as a desulphurizing unit than any of the other types in the paper.

MR. KUNIANSKY: Quite so. Someone might want to know why we use the box in Case No. 1, paragraph 15, that requires so much fused soda ash, especially in the light of our experience with the U-shaped ladle. We just unfortunately do not have the head room in our foundry to put in the other type of ladle.

F. HOLTBY[2]: May I ask the speaker how temperature measurements are made? I brought this point up last year as we found that in cast iron, where any material is being added that is taking sulphur out of the iron, that an optical pyrometer is completely in error.

We use an immersion type thermocouple just to get rid of the slag conditions. I am interested to know if you used an optical pyrometer on this.

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[1] Metallurgist, Deere & Co., Moline, Ill.

[2] Instructor in Foundry Practice, University of Minnesota, Minneapolis, Minn.

MR. KUNIANSKY: We use an optical. We do not claim that those measurements are accurate. Our hope is that the errors are uniformly wrong in these readings.

F. HOLTBY: I found in using a tungsten graphite thermocouple, an optical pyrometer and the platinum 10 per cent rhodium couple, if any desulphurization is done at all the optical immediately gives erroneous results because just that little bit of slag formation will change emissivity values. One can notice by eye that there is a distinct color change as desulphurization takes place, especially when in the ranges I am talking about.

CHAIRMAN PHILLIPS: We all recognize that, in using the optical, we have to have a clean metal surface on which to sight. If we do not, the optical is at a disadvantage.

MEMBER: The speaker made a little side issue of a melting rate of one of the cupolas. He mentioned that due to the nature of the stock, the cupola melting rate was decreased. Would he expand on that a little and describe the nature of that stock?

MR. KUNIANSKY: The nature of the stock does have considerable to do with how rapidly you can melt. We try to keep our blast pressure around 14 oz. due to the fact that we have a great deal of cotton mill scrap in our burden and also about 15 per cent stove plate scrap. When our pressures get too high, we find that our fluidity drops off and it is necessary in our super-deLavaud opertion to keep a steady and high fluidity. Fig. 6, showing a coke ratio of 11 to 1, is an operation on a balanced blast cupola.

MEMBER: I would like to ask Mr. Holtby another question about the optical pyrometer reading. Did you try that on a stream of iron emerging from a teapot spout?

F. HOLTBY: No. All our readings were taken either in the electric furnace or in the ladle, not on the stream. In this particular case we are working with, it is very difficult to get any reading out of the stream. Of course they are small furnaces.

MEMBER: You would not suspect any faulty reading there on the stream from the teapot spout?

F. HOLTBY: We know definitely that when we tap a high sulphur electric furnace iron that has manganese in it, causing a manganese sulphide reaction, the stream has an entirely different color than a stream of metal that is very low in sulphur. This quite often will fool a person very badly when measuring temperatures by eye.

There is an entirely different color change. One is very orange yellow color with high sulphur and the other is more toward the white yellow side. That is on the stream as it comes from the furnace.

For instance, we have caught that in one of our heats. One of the men working on it made a mistake and added the sulphur in the wrong heat. Right away we thought it was the wrong pouring temperature, just from the color of the stream. It is very definitely an orange yellow, where with low sulphurs below 0.15 per cent it is definitely a white yellow color.

Whether or not this affects the optical may be questioned, it should because the optical always must have a degree of correction.

We have a little test cupola we have just built, a 15-in. cupola, and we are going to start on production on test bars next week. We have taken off several heats, melting up some of this old cast iron that we have with high sulphur.

In the cast iron re-melted in the cupola up around 0.23 to 0.24 per cent sulphur that had the shot in it, the shot again came in the castings after they were run through the cupola, which to me again indicates that we are pouring iron which is not cold and that there is chemical reaction which is forming these little round balls. If it has gone through a cupola again and again, there must be some chemical reaction there that is causing that formation.